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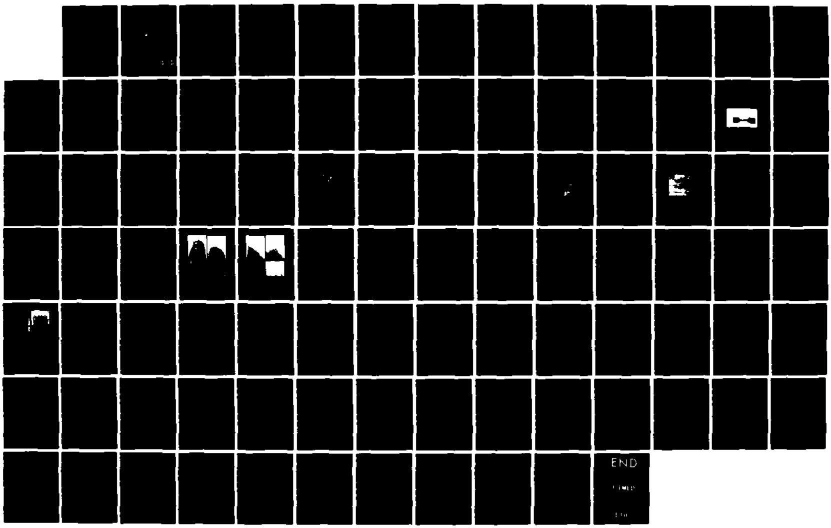

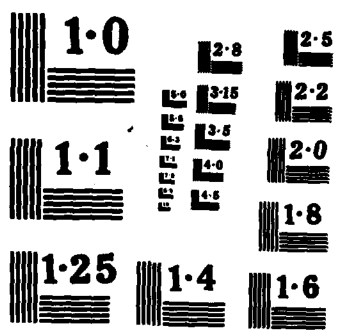
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NAVAL POSTGRADUATE SCHOOL

Monterey, California

THESIS

EFFECTS OF ANNEALING TREATMENTS ON SUPERPLASTICITY IN A THERMOMECHANICALLY PROCESSED ALUMINUM-10.2%MG-0.52%MN ALLOY

by

Alta Fern Stengel

December 1984

Thesis Advisor: T. R. McNelley

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20. ABSTRACT (Continue on reverse side if necessary and identify by block number)

This research follows previous thesis work by Becker and Mills on superplastic behavior of a warm rolled Al-10.2%Mg-0.52%Mn alloy. Elongations of up to 391% were reported by them for tension testing at 300°C and a strain rate of $1.4\times10^{-3}\ s^{-1}$. In this work, material was warm rolled at 300°C to 94% reduction and then subjected to one of five subsequent annealing treatments: 1 hour at 200°C, 10 hours

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at 200 °C, 1/2 hour at 250 °C, 1 hour at 250 °C, or 1/2 hour at 440 °C (to recrystallize the material). Tensiontesting at temperatures ranging from 300 °C to 425 °C was then conducted. Annealing below the rolling temperature enhances superplastic behavior when compared to the as-rolled condition. For example, material warm rolled at 300 °C, annealed for 1 hour at 200 °C and then tested at 300 °C with a strain rate of 5.6x10-3 s-1 gave a ductility of 572%. Annealing, however, for 1/2 hour at 440 °C results in a recrystallized structure which is stronger than the as-rolled condition and less ductile when tested at 300 °C.

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Effects of Annealing Treatments on Superplasticity in a Thermomechanically Processed Aluminum-10.2%Mg-0.52%Mn Alloy

by

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Submitted in partial fulfillment of the requirements for the degree of

MASTER OF SCIENCE IN MECHANICAL ENGINEERING

from the

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December 1984

Author: Alta F. Stengel

Approved by: T. R. McNelley, Thesis Advisor

P. J. Marto, Chairman, Department of Mechanical Engineering

John N. Dyer, Dean of Science and Engineering

ABSTRACT

This research follows previous thesis work by Becker and Mills on superplastic behavior of a warm rolled Al-10.2%Mg-0.52%Mn alloy. Elongations of up to 391% were reported ~~by them~~ for tension testing at 300°C and a strain rate of .0014 ~~$1.4x10^{-3}/s^{-1}$~~. In this work, material was warm rolled at 300°C to 94% reduction and then subjected to one of five subsequent annealing treatments: 1 hour at 200°C, 10 hours at 200°C, 1/2 hour at 250°C, 1 hour at 250°C, or 1/2 hour at 440°C (to recrystallize the material). Tension testing at temperatures ranging from 300°C to 425°C was then conducted. Annealing below the rolling temperature enhances superplastic behavior when compared to the as-rolled condition. For example, material warm rolled at 300°C, annealed for 1 hour at 200°C and then tested at 300°C with a strain rate of .0056 ~~$5.6x10^{-3}/s^{-1}$~~ gave a ductility of 572%. Annealing, however, for 1/2 hour at 440°C results in a recrystallized structure which is stronger than the as-rolled condition and less ductile when tested at 300°C.

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I. INTRODUCTION

A. SUPERPLASTIC BEHAVIOR

Serious study of superplastic behavior began following Underwood's review of superplasticity in 1962 [Ref. 1]. Since then, the term superplastic has generally meant high elongations (greater than 200%) at elevated temperatures and low strain rates. Superplastic behavior reflects characteristics of both the material microstructure, and the strain rate and temperature dependence of material strength. Five requirements for superplastic behavior are: (1) pronounced dependence of the flow stress on strain rate; (2) fine grain size; (3) thermally stable microstructure; (4) deformable second phase (if present); and (5) resistance to cavitation [Ref. 2] and [Ref. 3].

At elevated temperatures the flow stress (σ) may be related to strain rate ($\dot{\epsilon}$) by

$$\sigma = k\,\dot{\epsilon}^{m} \qquad \text{(eqn. 1.1)}$$

where m is the strain rate sensitivity, defined as:

$$m = \frac{d(\ln\sigma)}{d(\ln\dot{\epsilon})} \qquad \text{(eqn. 1.2)}$$

Equation 1.2 may be used to obtain m even though ln σ versus ln $\dot{\epsilon}$ is not linear over a wide range of strain rates, that is, m is a "local" value which applies only over a small range of strain rates. Values of m greater

than 0.3 are associated with superplastic behavior and increased material resistance to localized necking. Maximizing m gives the greatest superplastic response in the material. To achieve a large m typically requires a fine grain size less than 10um.

Two possible processes involved in superplastic behavior are (1) diffusional creep, i.e. Nabarro-Herring creep [Ref. 4] and (2) grain boundary sliding. The Nabarro-Herring diffusion creep model is not a completely adequate representation of superplastic behavior, however, it does illustrate the sensitivity of strain rate to grain size. Equation 1.3 is the Nabarro-Herring relationship:

$$\dot{\epsilon} = \frac{7b\sigma D}{kTd^2} \qquad \text{(eqn 1.3)}$$

where $\dot{\epsilon}$ is the strain rate, b the Burgers vector, σ the stress, D the diffusion coefficient, k a material constant, T the temperature and d the grain diameter. From this relationship it is clear that an increase in grain diameter d will mean a increase in stress σ if the strain rate $\dot{\epsilon}$ is held constant. If grain growth were to occur during a test, the material would "strain harden". Grain boundary sliding with diffusional accommodation is generally accepted as the flow process occurring during superplastic behavior and is associated with high m values in the strain rate sensitivity equation (eqn 1.1) mentioned earlier. A diffusion-accommodated flow model by Ashby and

Verrall [Ref. 5] proposes that polycrystalline material deformed at temperatures above $0.3T_m$ elongates by grain boundary sliding where grains switch their neighbors and remain essentially equiaxed rather then elongate when deforming superplastically. The rate of this diffusion accommodated flow may be limited by diffusion or by the capacity of grain boundaries to act as a sources or sinks for point defects, i.e. the interface reaction. The nature of this grain boundary is also described by Sherby [Ref. 6] as necessarily high-angle, because low-angle boundaries such as those comprising subgrain structures resulting from warm rolling do not slide readily under shearing stresses, as well as being poor sources or sinks for vacancies.

In order to prevent grain growth in superplastic forming, some form of grain boundary pinning is necessary [Ref. 2], [Ref. 7] and [Ref. 8]. A fine precipitate size will enhance the materials ability to resist grain growth as seen in the Zener-McLean relationship:

$$d \approx \frac{4}{3}\frac{r}{f} \qquad \text{(eqn 1.4)}$$

where d is the distance between pinning particles of radius r, and volume fraction f.

These grain boundary pinners should be deformable, otherwise cavities will form during rolling or forming processes. Certain brittle binary intermetallics tend to soften when they reach an inflection temperature T_i which

Petty [Ref. 9] reports as approximately 300 °C for intermetallics such as Al_8Mg_5 and $CuAl_2$, but around 500 °C for other intermetallics such as $FeAl_3$, $CrAl_7$,and Co_2Al_9.

Finally, Stowell's [Ref. 3] review of cavitation associates cavitation with second phase particles and inclusions. Cavitation may result from decohesion the of particle matrix interface, or plastic deformation during forming by stress-assisted vacancy diffusion. This may be produced by coarse constituent particles produced in casting or by precipitation of a brittle second phase. In order to avoid cavitation, precipitated particles and grain size should be kept as small as possible.

B. SUPERPLASTICITY IN ALUMINUM ALLOYS

Superplasticity in aluminum was first observed in eutectic and eutectoid alloys. These systems were thought necessary to achieve the fine grain size required for superplasticity. Presnyakov and Starikova [Ref. 10] reported superplastic behavior of eutectic Al-Cu (33% Cu) and proposed that such a high Cu-content was needed to obtain superplasticity in a cast Al-Cu Alloy. In addition, Holt and Backofen [Ref. 11] proposed that superplasticity in Al-Cu eutectic alloys requires a fine grain size to assist grain boundary shear, another possible rate controlling process.

Patton [Ref. 7] reported significant superplasticity in conventional high strength aluminum when grain size was refined to approximately 10um. High strength aluminum alloys such as 7075 and 7475 are made superplastic by solution treating at 482°C, annealing (overaging) at 400°C, warm rolling at approximately 200°C, and recrystallizing at 482°C prior to mechanical testing [Ref. 7]. For the 7475 alloy the best superplastic response was obtained at 516°C and a strain rate of $5 \times 10^{-4}\ s^{-1}$.

In Lloyd's [Ref. 2] review of superplasticity, magnesium and copper solute atoms are considered to interact with dislocations, preventing their movement and formation of low energy subgrains which would lower the driving force for recrystallization. Dispersoid forming elements such as zirconium or manganese were considered impurities that are capable of grain pinning but which also may form coarse, brittle particles undesirable because they crack during warm rolling and form initiation sites for cavitation.

In all cases there appears a need for fine recrystallized grains for superplastic behavior. Such a microstructure is achieved primarily by either cold or warm working the material then annealing just above the solvus for the strengthening component to recrystallize the microstructure.

C. HIGH MAGNESIUM-ALUMINUM ALLOY WORK AT THE NAVAL POSTGRADUATE SCHOOL

The materials engineering group at the Naval Postgraduate School has been studying Al-Mg alloys since 1976 when Ness [Ref. 12] initiated the investigation with research on an 18% Mg aluminum alloy. This material was mechanically worked in the two phase region of the phase diagram in an effort to obtain improved mechanical properties and grain refinement. As reported by Mondolfo [Ref. 13] and Becker [Ref. 14], an aluminum alloy with this high a magnesium content cracks easily during rolling and may exhibit very little ductility. Research with 7-12 percent magnesium aluminum alloys by Grandon [Ref. 15] showed that warm rolling below the solvus did not produce recrystallization. Therefore, the microstructure expected for Grandon's alloys would be low angle grain/subgrains boundaries with small misorientations. Through the preliminary work of Chesterman [Ref. 16] with 8% Mg aluminum alloys it was observed that recrystallization in a cold worked, cold worked then warm rolled, or just warm rolled 8% Mg aluminum alloy will only occur at temperatures at or above the solvus for Mg in the alloy. Warm rolling below the solvus of for Mg, with or without prior cold work, produced precipitation and recovery only. Chesterman concluded that warm working strengthens materials by mechanisms similar to cold work strengthening of solid

solution alloys and that these results obtained for 8% alloys could be applied to 10% alloys based on a review of the work of Grandon [Ref. 15] and Speed [Ref. 17].

Presently aluminum-magnesium alloys of 8-10% Mg processed at the Naval Postgraduate School are solution treated, upset forged on heated platens, annealed for 1 hour at 440 °C, oil quenched and then warm rolled at 300 °C following the procedures developed in the work of Johnson [Ref. 18] and Shirah [Ref. 19]. These alloy positions are indicated in the phase diagram of Figure 1.1 [Ref. 14]. Johnson [Ref. 18] concluded that the beta phase Al_8Mg_5 contributed towards achievement of a high strength, ductile material, and that at temperatures below the solvus precipitation during warm rolling prevented recrystallization of the alloy and hence facilitates strengthening in these alloys. However, as the temperature is raised above the solvus resolutioning of the Al_8Mg_5 occurs with corresponding loss in yield strength as well as recrystallization.

In addition, rolling below the solvus develops a fine dispersion of β and also breaks up grain boundary precipitates. Precipitation of the β at grain boundaries will cause susceptibility to stress-corrosion cracking when exposed to many environments at ordinary temperatures.

Finally, these high magnesium aluminum alloys also have a decreased density due to the increase in magnesium

content which provides a more desirable strength to weight ratio than unalloyed aluminum, which is important particularly in the aerospace industry.

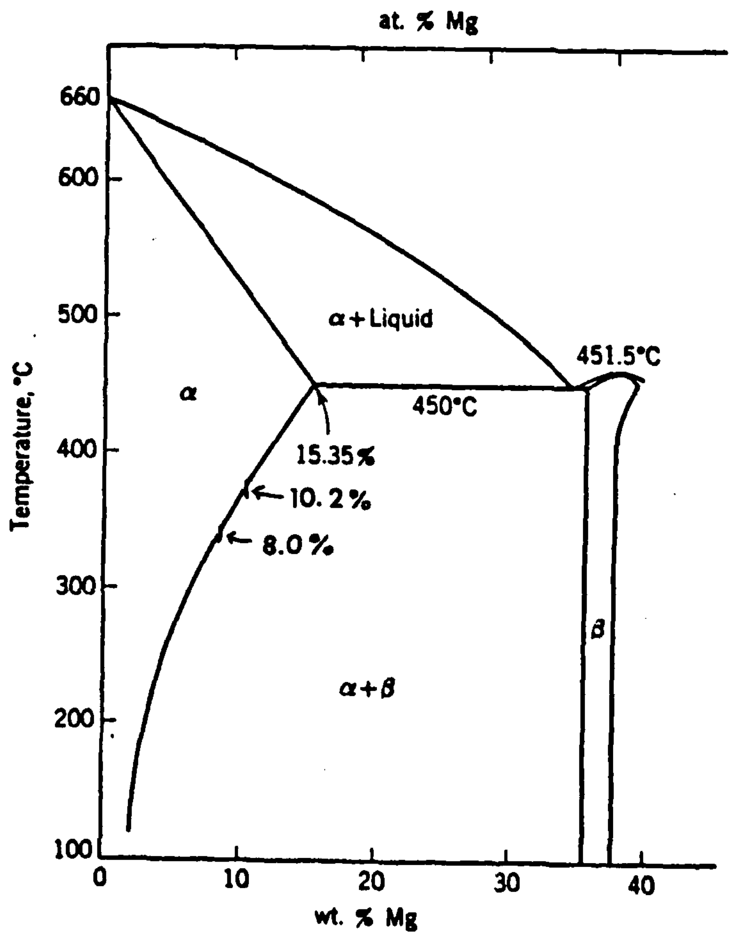


Figure 1.1 Partial Aluminum-Magnesium Phase Diagram.

II. BACKGROUND

A. PREVIOUS WORK

Recent research by Becker [Ref. 14] and Mills [Ref. 20], with the as-warm-rolled Al-10.2%Mg-0.52%Mn alloy of this thesis clearly demonstrated that thermo-mechanically processing of this alloy at 300°C produces a material which exhibits superplastic elongations when deformed near the rolling temperature (300°C). According to research by McNelley and Garg [Ref. 21], this as-rolled material appears to have a microstructure with fine subgrains as opposed to fine grains. This is a surprising result! Currently accepted theories suggest superplasticity should occur only in recrystallized structures with high angle grain boundaries, implying that a recovered structure in the material is detrimental to superplasticity. This data, then, suggests at least two possibilities: (1) another mechanism for superplasticity involving subgrain structures; (2) possible recrystallization during plastic deformation to provide a fine grain structure.

Additionally, McNelley and Garg [Ref. 21] annealed the as-rolled material at 300°C to see if the material was recrystallizing when the material was being heated prior to mechanical deformation. The microstructure of this annealed material also appeared to be recovered rather than

recrystallized. Becker [Ref. 14] subsequently annealed the as-rolled material for 1/2 hour and 10 hours at 300 °C prior to tensile testing at and below 300 °C. Also, Becker [Ref. 14] annealed another series of samples above the solvus, for 1/2 hour at 440 °C, to obtain recrystallization. These samples were tested as well with a view to comparison of recrystallized with as-rolled, or as-rolled and annealed material. Becker's results indicated enhanced ductility for the material annealed at 300 °C over material in the as-rolled condition which in turn was more ductile than the recrystallized material at this temperature. Since Becker's work, researchers at the ALCOA Technical Center [Ref. 22] have obtained x-ray crystallographic texture data for this alloy and again found results that suggest recovery rather than recrystallization when annealing is conducted at the rolling temperature (300 °C).

B. PURPOSE OF THESIS

The purpose of this thesis was to investigate characteristics of a thermo-mechanically processed Al-10.2%Mg-0.52%Mn alloy annealed below the rolling temperature prior to elevated temperature deformation.

Recrystallizing the as-rolled material prior to deformation is similar to current processing of aluminum alloys for superplastic forming. With this in mind,

elevated temperature testing from 300 °C to 425 °C was performed on material that had been warm rolled at 300 °C and recrystallized. This data, together with Becker's [Ref. 14] data at lower temperatures provided a baseline of comparison between recrystallized and as-rolled material. Having established this baseline of data between as-rolled and recrystallized material, further research on annealing below the rolling temperature was then conducted.

Experimental techniques and laboratory equipment for thermo-mechanical processing and elevated temperature tensile testing employed in this thesis are the same as those utilized by Becker [Ref. 14] and Mills [Ref. 20] in their research here at the Naval Postgraduate School.

Results and discussion of elevated temperature testing on recrystallized material and material annealed below the rolling temperature are presented together with optical micrographs of this alloy before and after mechanical testing.

III. EXPERIMENTAL PROCEDURE

A. MATERIAL PROCESSING

The material of this research was sectioned from the same ingot utilized by Becker [Ref. 14] and Mills [Ref. 20] in their work on superplastic aluminum alloys. The Al-10.2%Mg-0.52%Mn alloy was direct-chill cast at the ALCOA Technical Center [Ref. 18]. Composition details are provided in Table I [Ref. 20].

Table I

Alloy Composition (Weight Percent)

Serial Number	Si	Fe	Mn	Mg	Ti	Be
501300A	0.01	0.03	0.52	10.2	0.01	0.0002

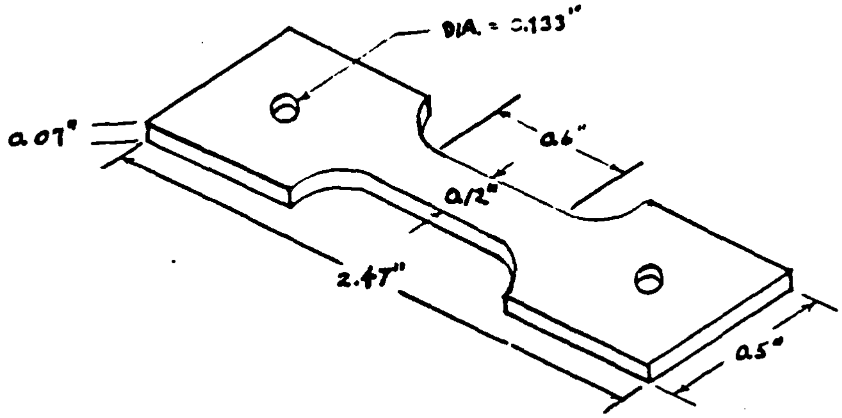
Billets of dimension 32mm x 32mm x 102mm (1.25in. x 1.25in. x 4.0in.) were sectioned from the as-cast ingot and then solution treated at 440 °C for 24 hours, upset forged at 440C, annealed at 440 °C for 1 hour and finally oil quenched. This portion of the processing was done by Mills [Ref. 20] according to procedures developed by Johnson [Ref. 18] and Becker [Ref. 14]. This hot working by upset forging reduced the billet by approximately 73%, equivalent to a strain of roughly 1.3 [Ref. 20].

Warm rolling of the billet was then done at 300 °C, with a further reduction in thickness of 94%, corresponding to a strain of 2.8. Blanks of dimension 63.5mm x 14.2mm x 1.8mm (2.46in. x 0.5in. x 0.07in.) were cut from the warm rolled material and formed into tensile specimens by endmilling [Ref. 14, 20]. Dimensions of the gage section of finished tensile specimens were approximately 3.0mm x 1.8mm x 15.2mm (0.12in. x 0.07in. x 0.6in., see Figure 3.1) [Ref. 20].

B. ANNEALING

Small groups of 8-9 tensile specimens were annealed in a salt bath for 1 hour at 200 °C, 10 hours at 200 °C, 1/2 hour at 250 °C, 1 hour at 250 °C, and 1/2 hour at 440 °C (above the solvus, to effect recrystallization) respectively. The annealing treatment differed from that done in previous work by Becker [Ref. 14], where the tensile specimens were placed on a heat sink in an air furnace.

The salt bath comprised approximately 2 liters of Cal Alloy #3010 Quenching Salt (temperature range 149 °C-593 °C) heated in a Lindberg Type 56622 crucible furnace. Fresh salt was melted in the furnace and the temperature allowed to stabilize overnight. Three Omega Type K Chromel-Alumel thermocouples connected to a Newport Laboratories Inc. digital readout were used to measure and monitor the



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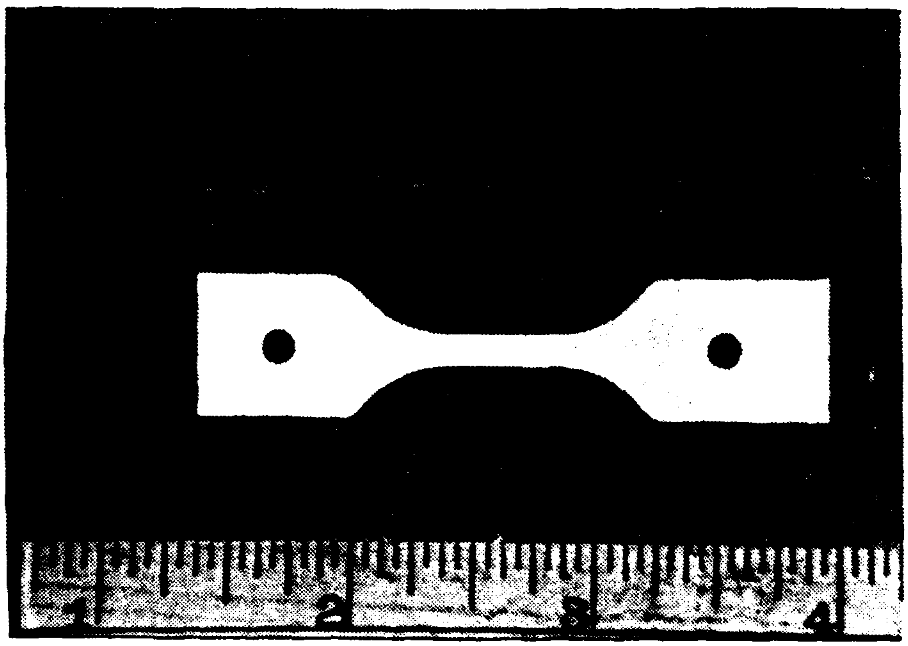


(b)

Figure 3.1 (a) Dimensions of the Tensile Test Specimen. (b) Photograph of Tensile Test Specimen.

annealing temperature of the salt bath. Temperatures were controlled within ± one degree Celsius throughout the salt bath, except very near the bath surface and positions immediately adjacent to the bottom and sides of the furnace.

The tensile specimens were suspended from a horizontal steel rod by a thin Nichrome wire so that they would be completely immersed in the salt bath and yet not touch the bottom. Eight centimeters (3 inches) of fiber glass insulation was carefully placed around the cover of the furnace to assist in maintaining a stable salt bath temperature. Annealed specimens were quickly removed from the salt bath and hung in a well-ventilated area to air cool. The tensile specimens were then carefully marked with a permanent ink to identify the annealing condition.

C. TENSILE TESTING

Elevated temperature testing was performed at 300°C, 325°C, 350°C, and 425°C with strain rates ranging from $1.39 \times 10^{-1}\ s^{-1}$ to $5.56 \times 10^{-5}\ s^{-1}$. An Instron tensile testing machine equipped with a Marshall clamshell furnace, as described by Mills [Ref. 20] and also by Becker [Ref. 14], was used to complete the testing sequence. The clamshell furnace and tension grips were preheated for 24 hours prior to starting the tensile tests at a given temperature. After loading the test specimen in the tension grips an

hour was allowed for the test specimen to become fully isothermal. Test specimen temperature was monitored with five thermocouples and remained nearly constant during all tests except at very low strain rates or upon attainment of high ductilities, where the bottom grip rod would begin to pull out of the furnace and the temperature of the rod would then drop as much as 5 to 10 degrees C. Otherwise, the tensile specimen temperature would remain constant within ± 2-3°C. Care was taken to follow exactly the experimental procedures of Mills [Ref. 20] and Becker [Ref. 14] so that experimental results would be comparable.

D. DATA REDUCTION

Tensile specimen ductility was determined by direct measurement of the test specimen before and after tension testing. Engineering stress and engineering plastic strain were calculated from the Instron strip chart recorder data and the initial measurements of the tensile specimen cross sectional area and gage length. A "floating slope" calculation similar to that of Mills [Ref. 20] was utilized to compensate for the inherent elasticity of the Instron machine, the elastic behavior of the tensile specimen itself and other factors such as grip tightening. The engineering stress and engineering plastic strain obtained were used to calculate true stress and true strain by use of the conventional formulas below for conversion of

engineering stress-engineering strain data to true stress and true strain:

$$\sigma_{true} = \sigma_{eng}(1+e) \qquad \text{(eqn 3.1)}$$

$$\epsilon_{true} = \ln(1+e) \qquad \text{(eqn 3.2)}$$

Where σ_{true} and ϵ_{true} are the true stress and strain, σ_{eng} is the engineering stress and e is the engineering strain.

E. COMPUTER PROGRAMS AND GRAPHING

Data reduction was accomplished on an IBM Personal Computer using Microsoft Basic and read into data files that were transmitted to the IBM 3033 at the Naval Postgraduate School via a telephone modem. All plotting was accomplished using EASYPLOT, an extremely user friendly program package which outputs DISSPLA plots on a Versatec plotter at the Naval Postgraduate School Computer Center. Basic programs used to reduce the data are included in Appendix B.

F. METALLOGRAPHY

Portions from the gage region of tested tensile specimens for strain rates of 5.56×10^{-2}, 5.56×10^{-3}, and 5.56×10^{-4} per second, annealing conditions of 1 hour at 200°C and 1/2 hour at 440°C, and the previously specified range of tension test temperatures were cold mounted with an epoxy resin compound for metallographic examination in a manner similar to that described by Mills [Ref. 20]. In

addition, untested tensile specimens of as-rolled material, material annealed for 1 hour at 200 °C, and also material annealed for 1/2 hour at 440 °C were cold mounted for metallographic examination. All mounted tensile specimen surfaces were polished for optical microscopy using the techniques described by Becker [Ref. 14] and Mills [Ref. 20]. The polished tensile specimens were either etched for 60 seconds with a Graf-Sargent solution or etched for 3 minutes with Keller's solution, and examined with a Zeiss Universal Optical Microscope. Photomicrographs were taken using Kodak 35mm Technical Pan Film 2415 and Kodak 35mm Panatomic X film.

IV. RESULTS AND DISCUSSION

A. STARTING STRUCTURE OF THE AS-ROLLED MATERIAL

Thermomechanical processing of the Al-10.2%Mg-0.52%Mn alloy began with all of the magnesium in solution at the outset of warm rolling, the result of prior solution treatment and quenching. Using the same rolling procedure employed in this research McNelley and Garg [Ref. 21] found precipitation of intermetallic ß, Al_8Mg_5, at rolling strains from 0.6 to 1.0 at 300°C [Ref. 21]. Thereafter, for increased true rolling strain the content of magnesium in solid solution leveled off at 7.2%, a reflection of the approximate solubility limit of magnesium in aluminum at this temperature [Ref. 21]. The aluminum-magnesium phase diagram predicts still smaller percentages of magnesium remaining in solution when rolling at lower temperatures due to the decreased solubility of magnesium in aluminum. Precipitation of the intermetallic at room temperature does not occur to a significant extent because low temperatures severely limit diffusion. Figure 4.1 illustrates the precipitation of magnesium during warm rolling and the corresponding increase in hardness with the increase of true rolling strain [Ref. 21].

Warm rolling the material at 300°C to 94% reduction (2.8 true strain) produces a dislocation substructure with

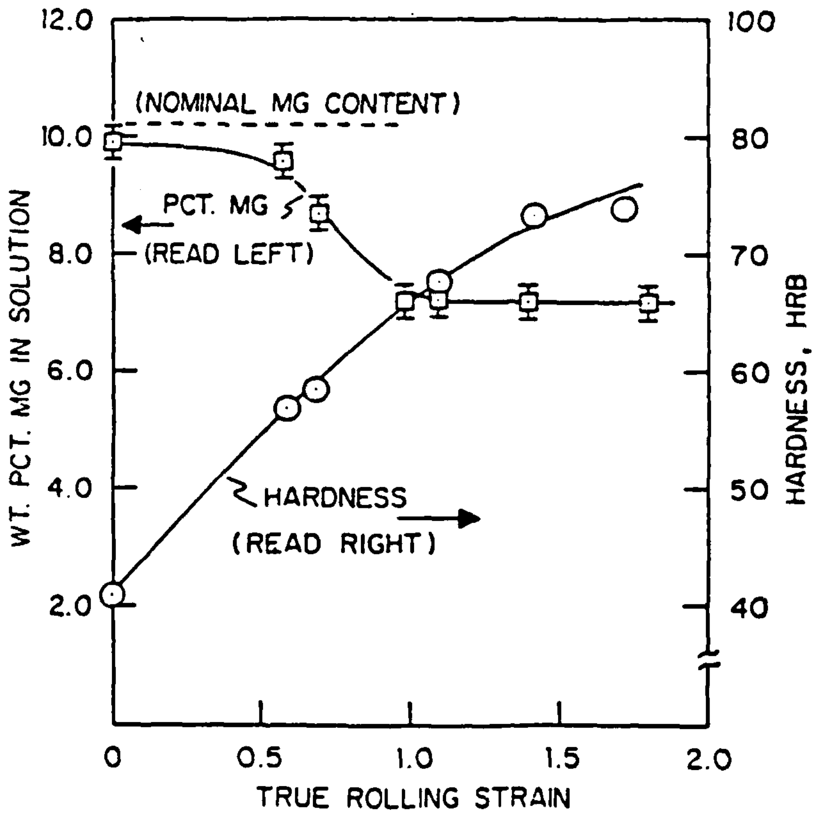


Figure 4.1 The Increase in Hardness and the Precipitation of Mg from Solution During Warm Rolling. Data from McNelley and Garg [Ref. 21].

precipitation of intermetallic β likely occurring at dislocation nodes according to McNelley and Garg [Ref. 21]. This uniform distribution of refined intermetallic β particles is shown in Figure 4.2 [Ref. 21].

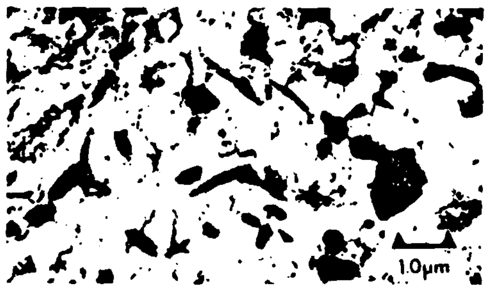


Figure 4.2 The Precipitated β Near the Foil Edge, Showing the β in Good Contrast and Uniformly Distributed as Fine Particles. Transmission Electron Micrograph from Ref. 21.

Although the precipitate dispersion may provide some strengthening, it is thought secondary to the effects of the remaining magnesium in solid solution and the dislocation structure produced by the warm rolling.

Previous work by Becker as well as Mills has associated this particular microstructure with high strength and good ductility at ambient temperature [Refs. 14, 20].

B. RATIONAL BEHIND ANNEALING BELOW THE ROLLING TEMPERATURE

As noted previously, the magnesium content of the solid solution for rolling done at 300 °C, is about 7%, the

equilibrium solubility for magnesium at this temperature. Annealing at temperatures below this would result in further precipitation. The solubility limit of magnesium in aluminum at 200 °C drops to 3.6%. This more than doubles the amount of magnesium precipitated from solution. By eqn 1.4, the additional precipitation may stabilize the grain size by deposition of fine β particles on the subgrain structure, which would help to pin grain boundaries and prevent grain growth. X-ray texture data from ALCOA Technical Center [Ref. 22] shows the structure after warm rolling at 300 °C, and after annealing for times of 1 and 10 hour(s) at 300 °C as well to be recovered rather than recrystallized. From this data it was inferred that annealing at temperatures below 300 °C would also imply a recovered material structure.

Figures 4.3, 4.4 and 4.5 show the difference in mechanical fibering of as-rolled material, for material annealed for 1 hour at 200 °C after warm rolling, and also material recrystallized by annealing for 1/2 hour at 440 °C after warm rolling. Notice that there appears to be little difference between the as-rolled material, Figure 4.3, and the material annealed for 1 hour at 200 °C, Figure 4.4. This is as expected since annealing at 200 °C would reasonably produce a very fine precipitation likely not observable with light microscopy, with only little growth in the intermetallic already precipitated. Additionally

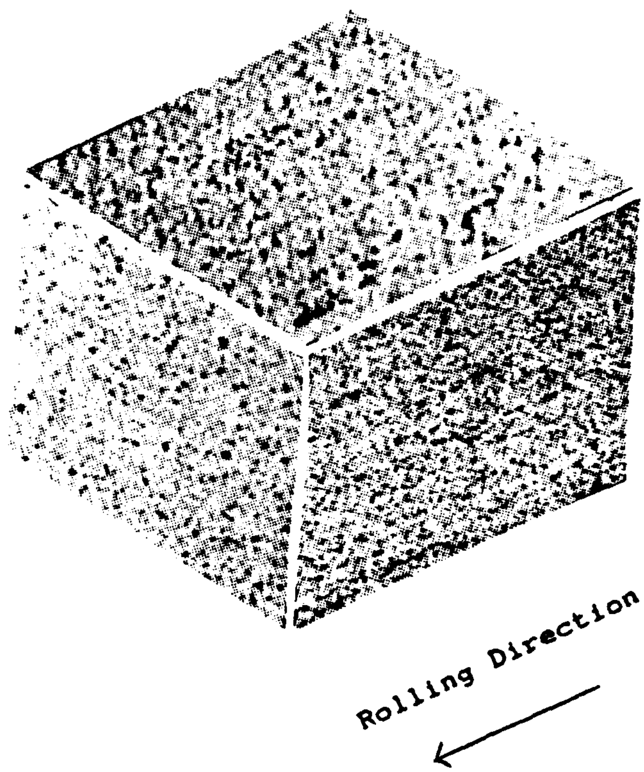


Figure 4.3 Tri-Planar Photomicrograph of Al-10.2%Mg-0.52%Mn Alloy in the As-Rolled Condition, Graf-Sargent Etch, x64.

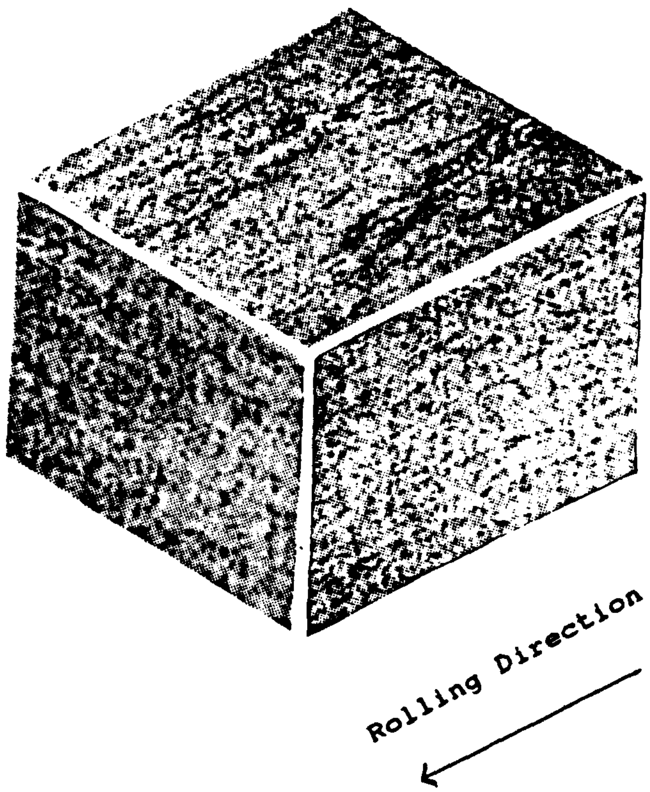


Figure 4.4 Tri-Planar Photomicrograph of Al-10.2%Mg-0.52%Mn Alloy Annealed for 1 Hour at 200°C. Notice That There is Little Difference in Precipitation of the Intermetallic Between This Photomicrograph and Figure 4.4, Graf-Sargent Etch, x64.

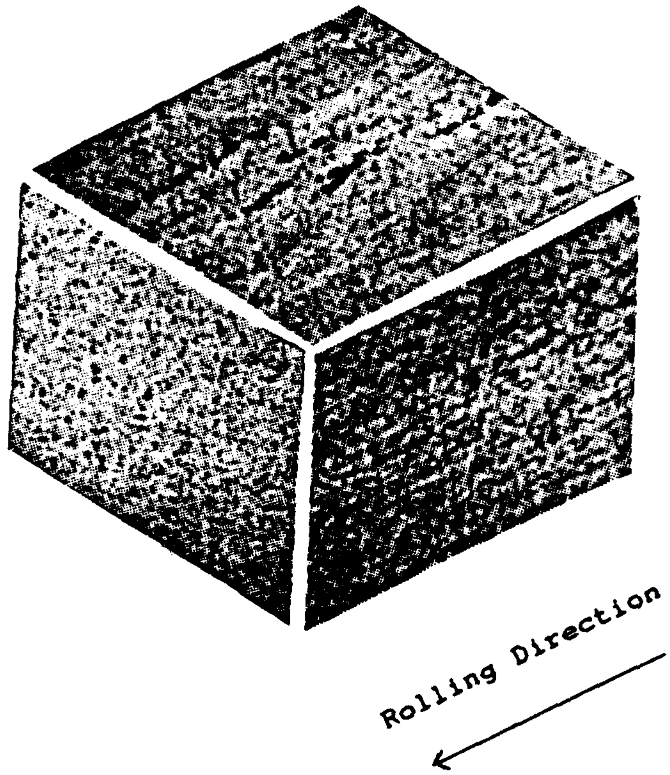


Figure 4.5 Photomicrograph of Al-10.2%Mg-0.52%Mn Alloy Recrystallized by Annealing Above the Solvus for 1/2 Hour at 440°C, Graf-Sargent Etch, x64.

the beta is actually decorated by the Graf-Sargent etch and is likely much smaller than shown. Figure 4.5 is of material recrystallized by annealing above the solvus for 1/2 hour at 440 °C. The etched intermetallic is a little coarser than Figures 4.3 and 4.4 but not significantly more. Figure 4.6 is of recrystallized material at roughly 5 times higher magnification than that of the photomicrograph in Figure 4.5. Grain size for the recrystallized material is on the order of 8um. The size of the grains in the as-rolled, and as-rolled and annealed materials could not be determined with light microscopy using the etching solutions of this thesis. This supports the ALCOA Technical Center's texture data indicating a recovered structure for this material.

Data from recrystallized material at temperatures ranging from 300 °C to 425 °C together with that of Mills [Ref. 20] for as-rolled material and also that of Becker [Ref. 14] for recrystallized material tested below 300 °C demonstrates superior ductility of the as-rolled material in the temperature range between 150 °C and 325 °C. Figure 4.7 clearly shows this result by plotting ductility vs temperature at a strain rate of $5.56 \times 10^{-3}\ s^{-1}$ for temperatures from 20 °C to 425 °C. The as-rolled condition is also weaker than the recrystallized material when tested at 300 °C (Figure 4.8) and exhibits enhanced superplastic

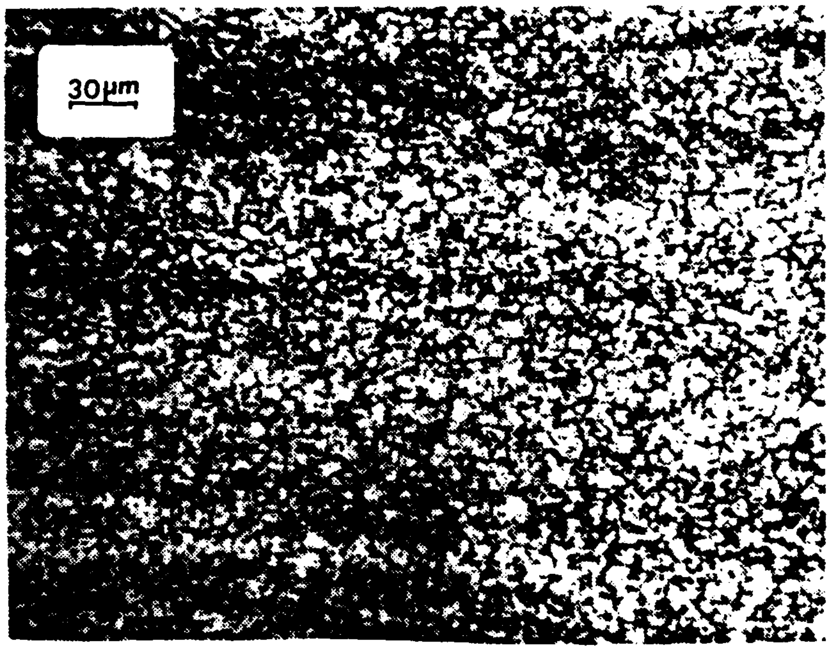


Figure 4.6 Photomicrograph of Al-10.2%Mg-0.52%Mn Alloy Recrystallized by Annealing Above the Solvus for 1/2 Hour at 440 °C. Grain Size is Approximately 8μm, Keller's Etch, x320.

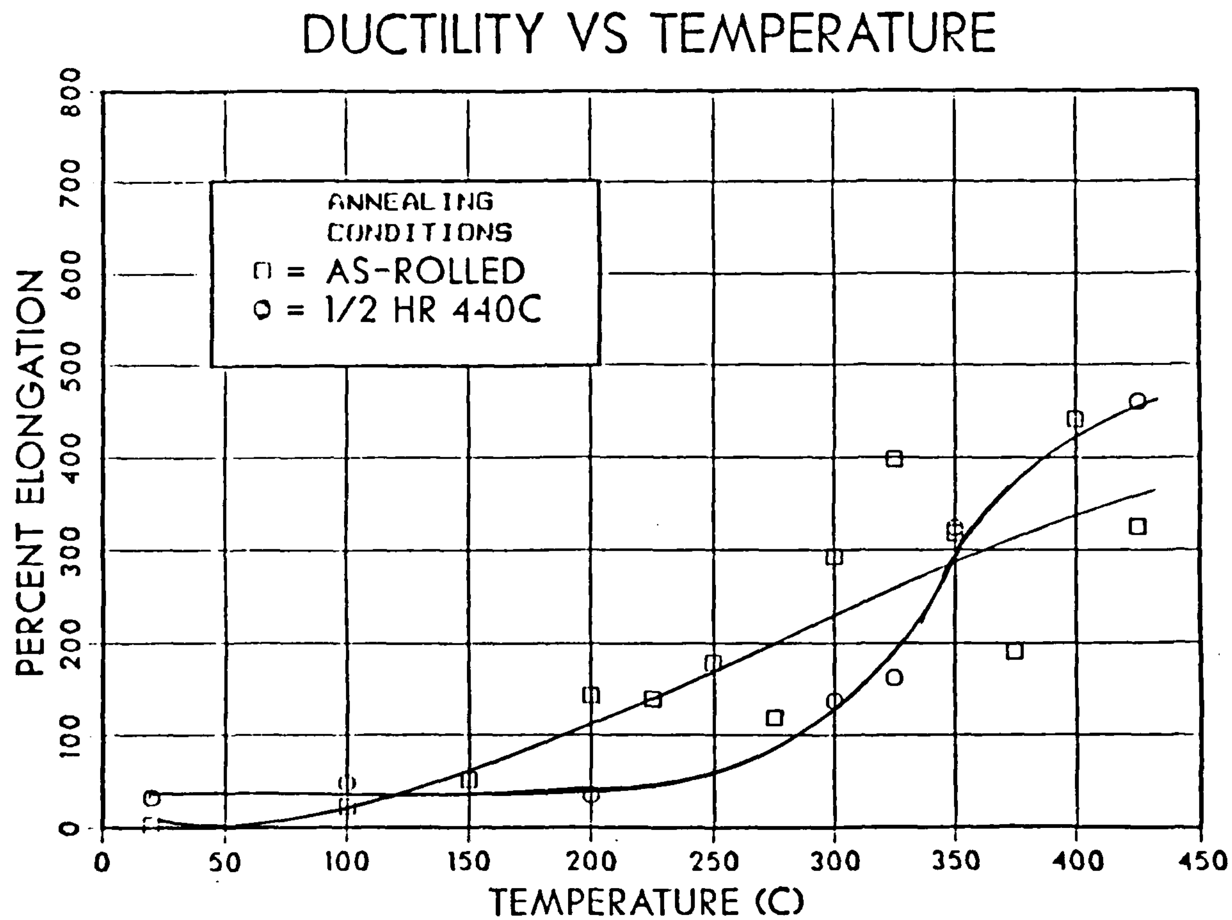


Figure 4.7 Ductility vs Temperature for Tensile Tests on Al-10.2%Mg-.52%Mn at $5.56 \times 10^{-3}\ s^{-1}$. Note the Enhanced Superplastic Behavior of the As-Rolled Material between 150C and 300C. As-Rolled Material Processed at 300C is recovered while Material Annealed for 1/2 Hour at 440C is Recrystallized.

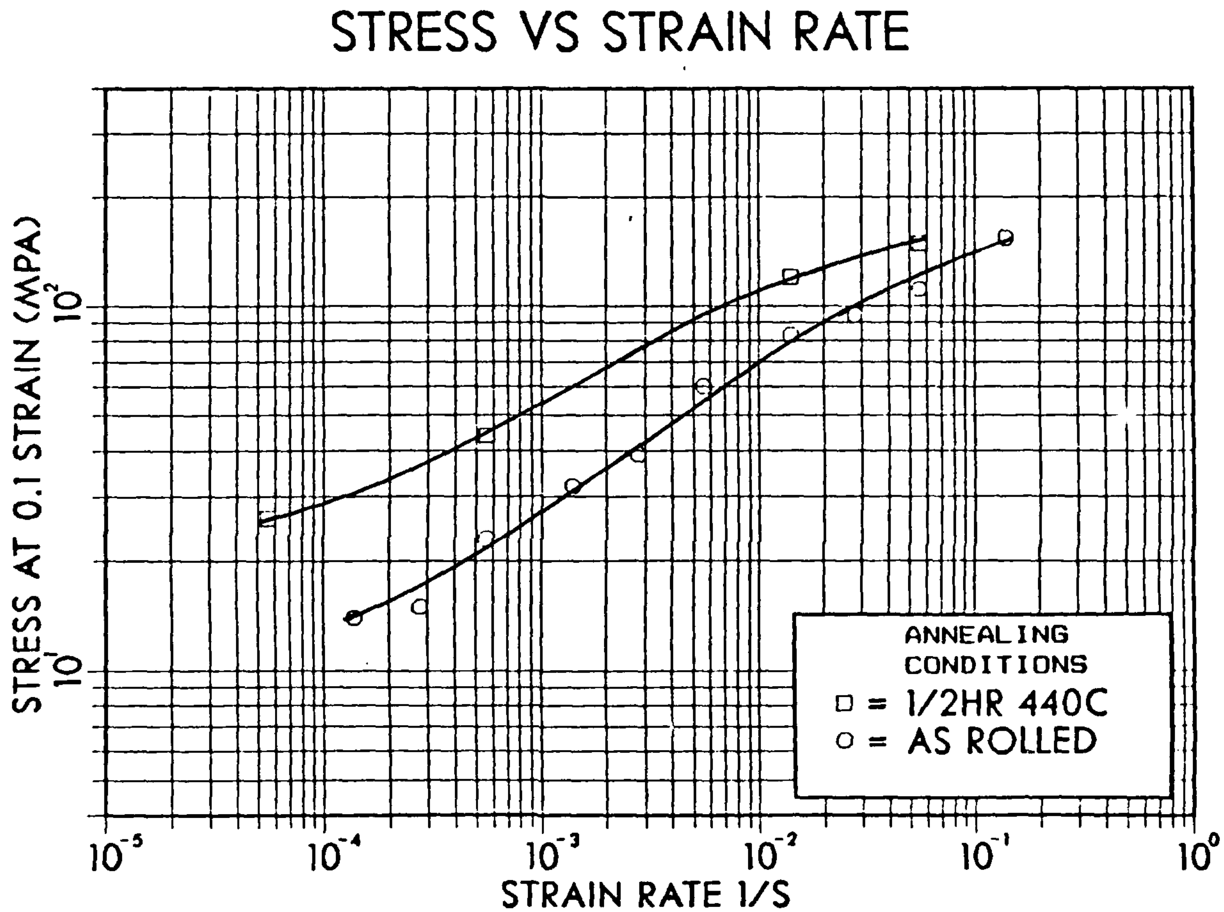


Figure 4.8 True Stress at 0.1 Strain vs Strain Rate for Tensile Tests on Al-10.2%Mg-0.52%Mn Conducted at 300C. The As-Rolled Condition is Weaker than Material Recrystallized by Annealing 1/2 Hour at 440C.

behavior compared to the recrystallized material, Figure 4.9. In Figure 4.8, the flow stress at 0.1 strain is plotted versus strain rate for the two processing conditions of interest. The data for the warm rolled material indicates both a lower strength at any strain rate in the range evaluated and also a larger value of $m = d\log\sigma/d\log\dot{\epsilon}$; $m \approx 0.45$ for the rolled material, and the data suggest a sigmoidal relation, while $m \approx 0.3$ for the material recrystallized prior to elevated temperature testing. The sigmoidal form of the stress-strain rate data for the rolled material is consistent with the ductility data of Figure 4.9; there, a maximum ductility of approximately 400% is seen at a strain rate of $\sim 2\times10^{-3}\ s^{-1}$, about the strain rate of maximum slope in the data of Figure 4.8. It should be recognized that the value of m is determined from data at a strain $\epsilon = 0.1$ and microstructural changes at greater strains could affect the value of m and through it the final ductility of the material. In any case, this is a surprising result! Current theory as discussed in the introduction and background of this thesis predicts superplasticity only in fine grained, recrystallized aluminum magnesium alloys with high angle grain boundaries. Test results for the recrystallized material between 300 °C and 425 °C agree with current theory when the test temperature is above the solvus. At 300 °C, for a given strain rate, the stress at 0.1 strain for

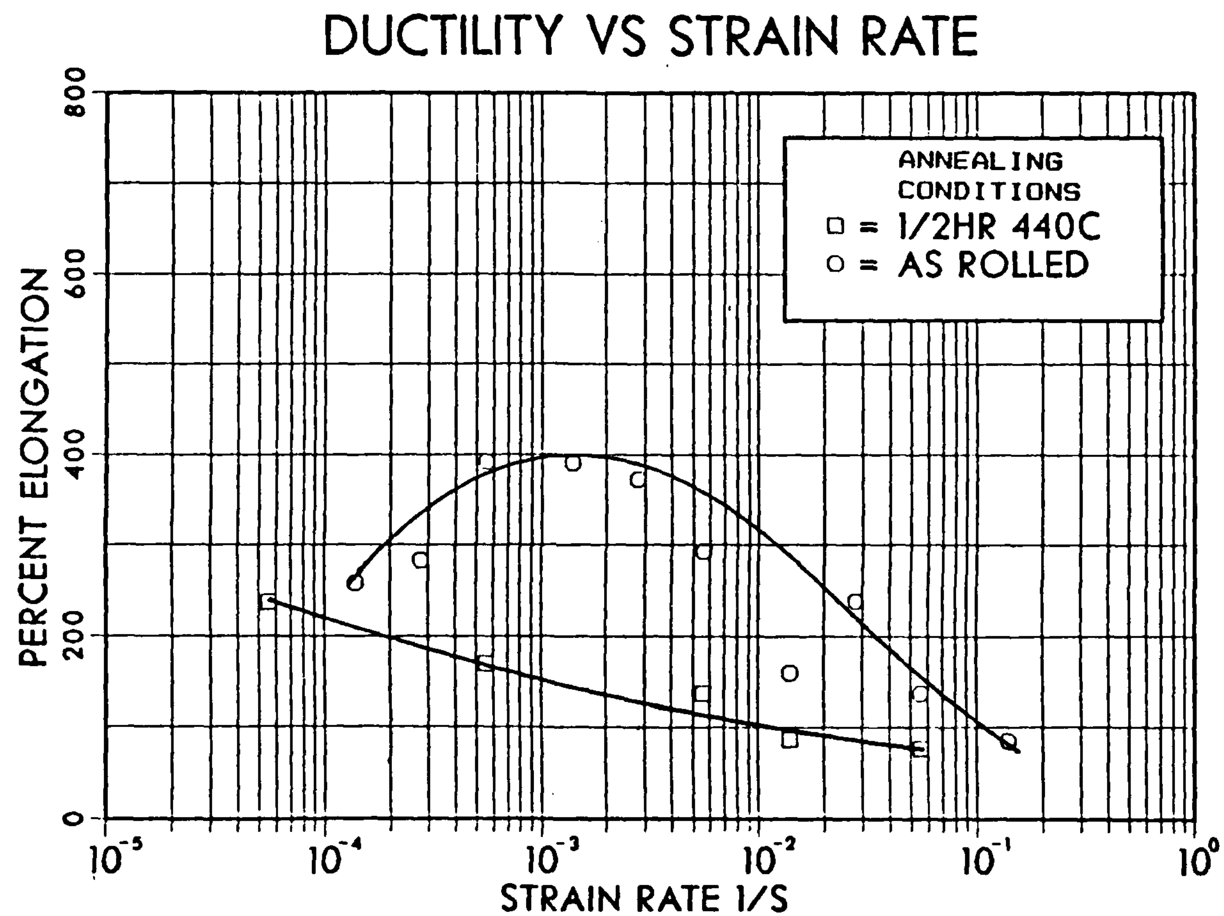


Figure 4.9 Ductility vs Strain Rate for Tensile Tests on Al-10.2%Mg-0.52%Mn Conducted at 300C. Annealing Conditions: Fully Recrystallized after Warm Rolling, and As-Rolled at 300C.

recrystallized material is higher than that for the as-rolled material. However, as test temperature increases, the strength of the recrystallized material decreases more rapidly than that of the as-rolled material and for temperatures at and above the solvus the two conditions behave in a similar manner. This would be anticipated as the as-rolled material would be expected to recrystallize as well upon initial heating above the solvus temperature. Most significantly, the recrystallized material tested at 300 °C shows no superplastic behavior even at low strain rates (ductility is 170%), but the recrystallized material is very superplastic i.e. 556% at a strain rate of $5.56x10^{-4}$ s^{-1}, when tested at 425 °C, again, in good agreement with reported behavior of other superplastic Al-alloys [Refs. 2, 7]. Figures 4.10 and 4.11 show the microstructure of this recrystallized material when tested at 425 °C and 300 °C respectively. Note that superplastically deformed, recrystallized material in Figure 4.10 has elongated grains when tested at a low strain rate $5.56x10^{-4}$ s^{-1}, and more cavitation is apparent than in the same material tested at a high strain rate, $5.56x10^{-2}$ s^{-1}.

As noted previously, the structure of the warm rolled material consists of recovered subgrains and intermetallic β precipitates as opposed to a recrystallized, fine grains. If the subgrain structure persists during plastic

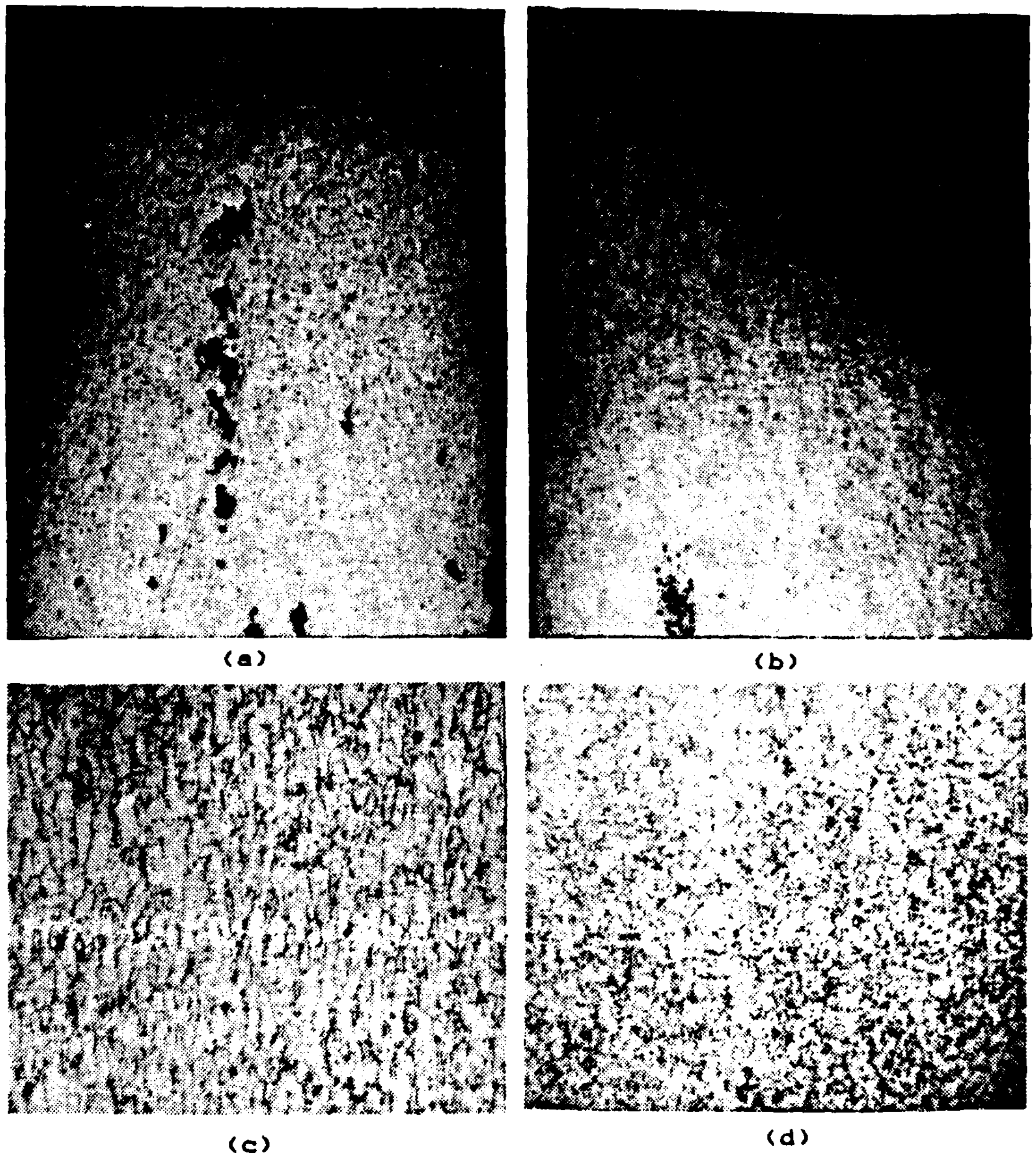


Figure 4.10 Recrystallized Material Tested at 425 °C at a Strain Rate of 5.56×10^{-4} (a) & (c), and 5.56×10^{-2} (b) & (d). Note Elongated Grains in (c) at the Lower Strain Rate as Opposed to the More Equiaxed Grains in (d) at the Higher Strain Rate. Elongation for (a) & (c) is 556%, and 327% for (b) & (d). Magnification for (a) & (b) is x64 and for (c) & (d) is x250, Graf-Sargent Etch.

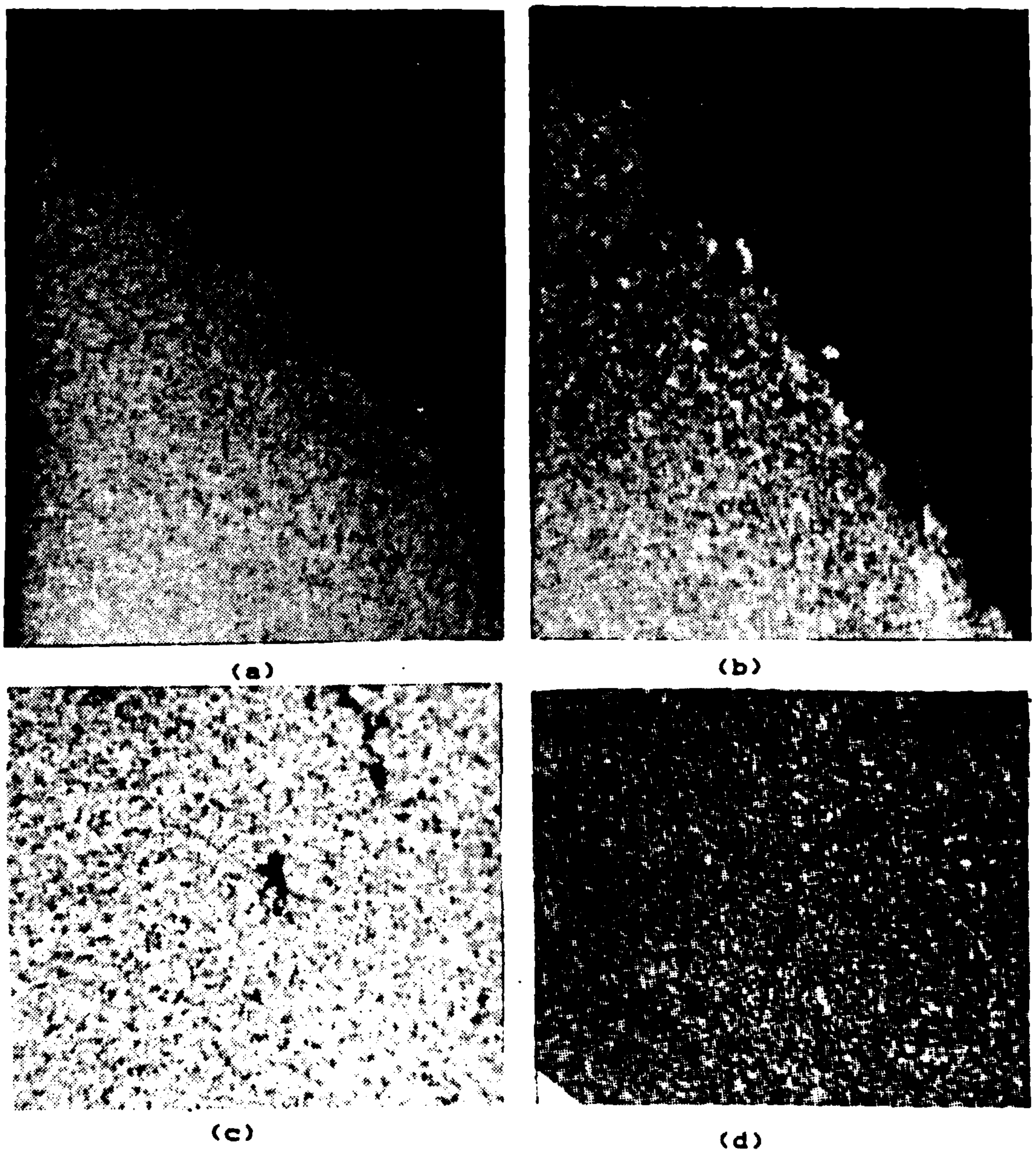


Figure 4.11 Recrystallized Material Tested at 300 °C at a Strain Rate of 5.56×10^{-4} (a) & (c), and 5.56×10^{-2} (b) & (d). Note Finer Grains in (b) & (d) than Figure 4.7 (b) & (d) and Much Less Cavitation, Also Recrystallized Material is Not Superplastic, Elongation for (a) & (c) is 170% and for (b) & (d) is 78%. Magnification for (a) & (b) is x64 and for (c) & (d) is x250, Graf-Sargent Etch.

deformation at 300 °C, then current theories are unable to account for observation of superplastic behavior under such conditions. Annealing at 300 °C results in further recovery with little apparent recrystallization taking place under such static conditions. One would not expect such annealing as results during heating prior to stress-strain testing to lead to recrystallization. It is possible, however, that recrystallization may occur as a result of plastic deformation, i.e. under dynamic conditions. Such recrystallization would then confer the fine, high angle boundaries generally though necessary for the elongations observed. Further research on materials deformed various amounts under such conditions, including microscopy of the deformed samples, would be necessary to test this. In the absence of such recrystallization, an alternative theory for superplastic flow would be necessary.

C. EFFECTS OF ANNEALING BELOW THE ROLLING TEMPERATURE

Extensive mechanical testing was performed at 300 °C for the following reasons: Becker's [Ref. 14] and later Mill's [Ref. 20] work indicated that this temperature was sufficient for superplastic behavior in as-rolled material; this temperature is also below the solvus and will not recrystallize the material under static annealing conditions; and, being below the solvus where recrystallization doesn't occur, one would not expect to

develop intergranular precipitates with attendant corrosion and stress-corrosion susceptibility.

Tables II through IX present the test data for material annealed below the rolling temperature and also for the recrystallized material. Overall, the material annealed at 200 °C displayed better superplastic behavior when compared to material annealed at 250 °C. The 1 and 10 hour annealing treatments at 200 °C did not differ significantly in results, and hence the results for one hour annealing at 200 °C will be presented exclusively here. Figure 4.12 shows true stress at 0.1 strain versus strain rate for three conditions. As seen from this plot, material annealed for 1/2 hour at 440 °C (which results in a recrystallized structure) is stronger than the as-rolled material as has been noted. Annealing for 1 hour at 200 °C results in lower strength when compared to as-rolled material. The stronger,recrystallized material has poorer ductility than the as-rolled material while annealing the material for 1 hour at 200 °C enhances the superplastic elongation. Figure 4.13 shows this enhancement of ductility in the annealed material. As noted previous, $m \approx 0.3$ for the recrystallized material while the maximum value of m determined form the data for as-rolled material is 0.45. It is difficult to state that m is still larger for the material rolled and then annealed at 200 °C given the inherent scatter in such data. However, the enhanced

Table II

Mechanical Test Data
for Warm Rolled
Al-10.2%Mg-0.52%Mn Alloy
Annealed for 1 Hour at 200C;
Tests Conducted at 300C

Strain Rate 1/s	UTS Mpa	True Stress at 0.1 Plastic Strain Mpa	Ductility %
1.39×10^{-1}	112.7	122.0	248
5.56×10^{-2}	98.1	108.0	301
1.39×10^{-2}	60.2	66.0	401
5.56×10^{-3}	44.7	49.0	572
1.39×10^{-3}	26.4	29.0	445
5.56×10^{-4}	18.5	20.0	437
1.39×10^{-4}	11.3	10.0	350

Table III

Mechanical Test Data
for Warm Rolled
Al-10.2%Mg-0.52%Mn Alloy
Annealed for 10 Hours at 200C;
Tests Conducted at 300C

Strain Rate 1/s	UTS Mpa	True Stress at 0.1 Plastic Strain Mpa	Ductility %
1.39×10^{-1}	107.5	118.0	197
5.56×10^{-2}	102.2	112.0	291
1.39×10^{-2}	60.1	60.0	396
5.56×10^{-3}	40.9	44.0	413
1.39×10^{-3}	25.5	27.0	679
5.56×10^{-4}	17.5	18.0	417
1.39×10^{-4}	15.7	16.0	255

Table IV

Mechanical Test Data
for Warm Rolled
Al-10.2%Mg-0.52%Mn Alloy
Annealed for 1/2 Hour at 250C;
Tests Conducted at 300C

Strain Rate 1/s	UTS Mpa	True Stress at 0.1 Plastic Strain Mpa	Ductility %
1.39×10^{-1}	106.3	138.0	156
5.56×10^{-2}	98.5	113.0	162
1.39×10^{-2}	57.9	62.0	280
5.56×10^{-3}	41.5	42.0	347
1.39×10^{-3}	27.0	29.0	407
5.56×10^{-4}	17.1	18.0	349
1.39×10^{-4}	12.6	15.0	215

Table V

Mechanical Test Data
for Warm Rolled
Al-10.2%Mg-0.52%Mn Alloy
Annealed for 1 Hour at 250C;
Tests Conducted at 300C

Strain Rate 1/s	UTS Mpa	0.1 Plastic Strain Mpa	Ductility %
1.39×10^{-1}	126.1	138.0	107
5.56×10^{-2}	103.9	113.0	157
1.39×10^{-2}	56.6	62.0	344
5.56×10^{-3}	39.7	42.0	543
1.39×10^{-3}	27.1	29.0	342
5.56×10^{-4}	16.6	18.0	409
1.39×10^{-4}	13.3	15.0	315

Table VI

Mechanical Test Data
for Warm Rolled
Al-10.2%Mg-0.52%Mn Alloy
Annealed for 1/2 Hour at 440C;
Tests Conducted at 300C

Strain Rate 1/s	UTS Mpa	True Stress at 0.1 Plastic Strain Mpa	Ductility %
5.56×10^{-2}	139.0	149.0	78
1.39×10^{-2}	114.0	120.0	87
5.56×10^{-4}	40.7	44.0	170
5.56×10^{-5}	23.5	26.0	237

Table VII

Mechanical Test Data
for Warm Rolled
Al-10.2%Mg-0.52%Mn Alloy
Annealed for 1/2 Hour at 440C;
Tests Conducted at 325C

Strain Rate 1/s	UTS Mpa	True Stress at 0.1 Plastic Strain Mpa	Ductility %
5.56×10^{-2}	108.0	114.0	107
5.56×10^{-3}	63.5	67.0	163
5.56×10^{-4}	24.3	26.0	413

Table VIII

Mechanical Test Data
for Warm Rolled
Al-10.2%Mg-0.52%Mn Alloy
Annealed for 1/2 Hour at 440C;
Tests Conducted at 350C

Strain Rate 1/s	UTS Mpa	True Stress at 0.1 Plastic Strain Mpa	Ductility %
$5.56x10^{-2}$	100.4	99.0	139
$5.56x10^{-3}$	14.0	16.0	325
$5.56x10^{-4}$	4.4	5.0	540

Table IX

Mechanical Test Data
for Warm Rolled
Al-10.2%Mg-0.52%Mn Alloy
Annealed for 1/2 Hour at 440C;
Tests Conducted at 425C

Strain Rate 1/s	UTS Mpa	True Stress at 0.1 Plastic Strain Mpa	Ductility %
$5.56x10^{-2}$	38.1	40.0	327
$5.56x10^{-3}$	14.2	13.0	460
$5.56x10^{-4}$	5.3	5.0	556

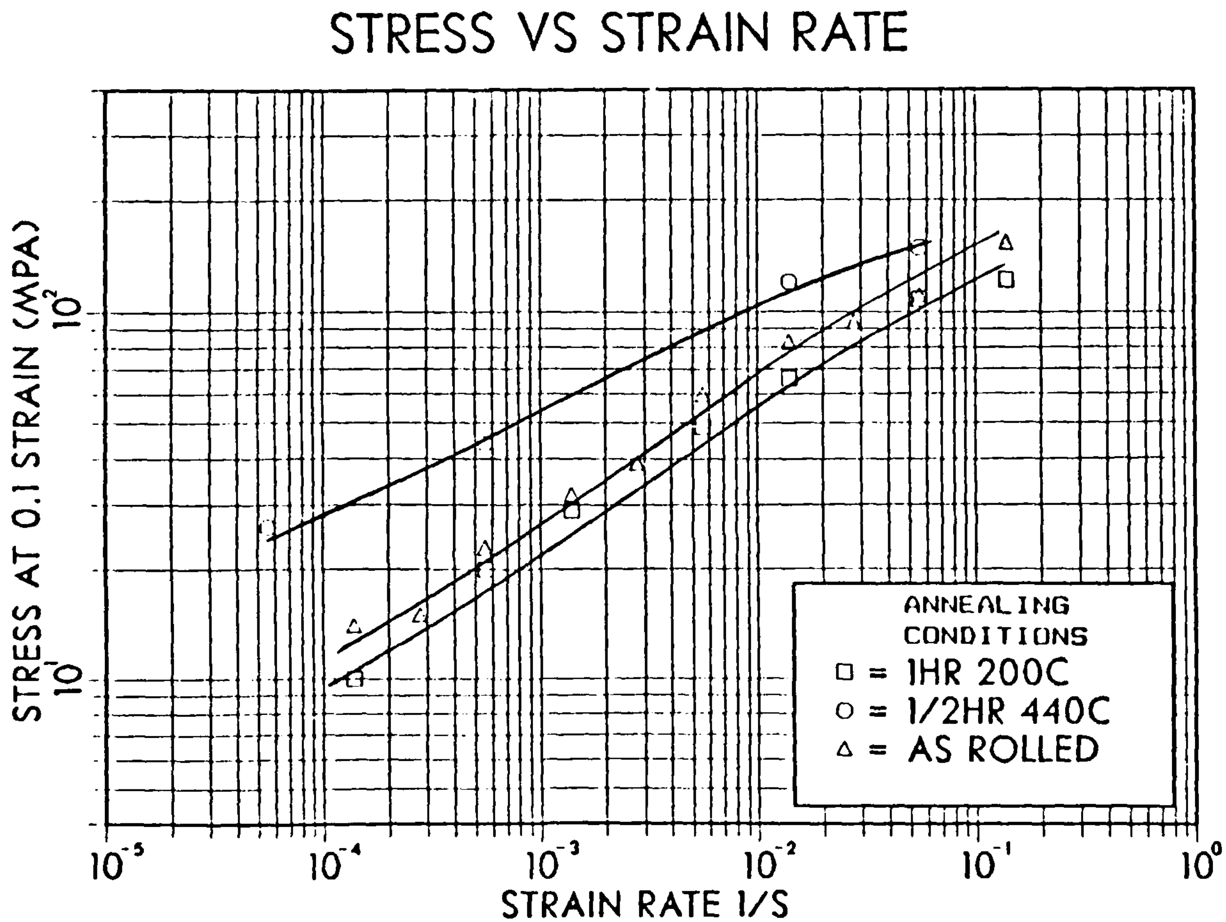


Figure 4.12 True Stress at 0.1 Strain vs Strain Rate for Tensile Tests on Al-10.2%Mg-0.52%Mn Conducted at 300C. Annealing for 1/2 Hour at 440C Results in a Recrystallized Structure Which is Stronger than the As-Rolled Condition While Annealing for 1 Hour at 200C Results in Lower Strength when Compared to As-Rolled Material.

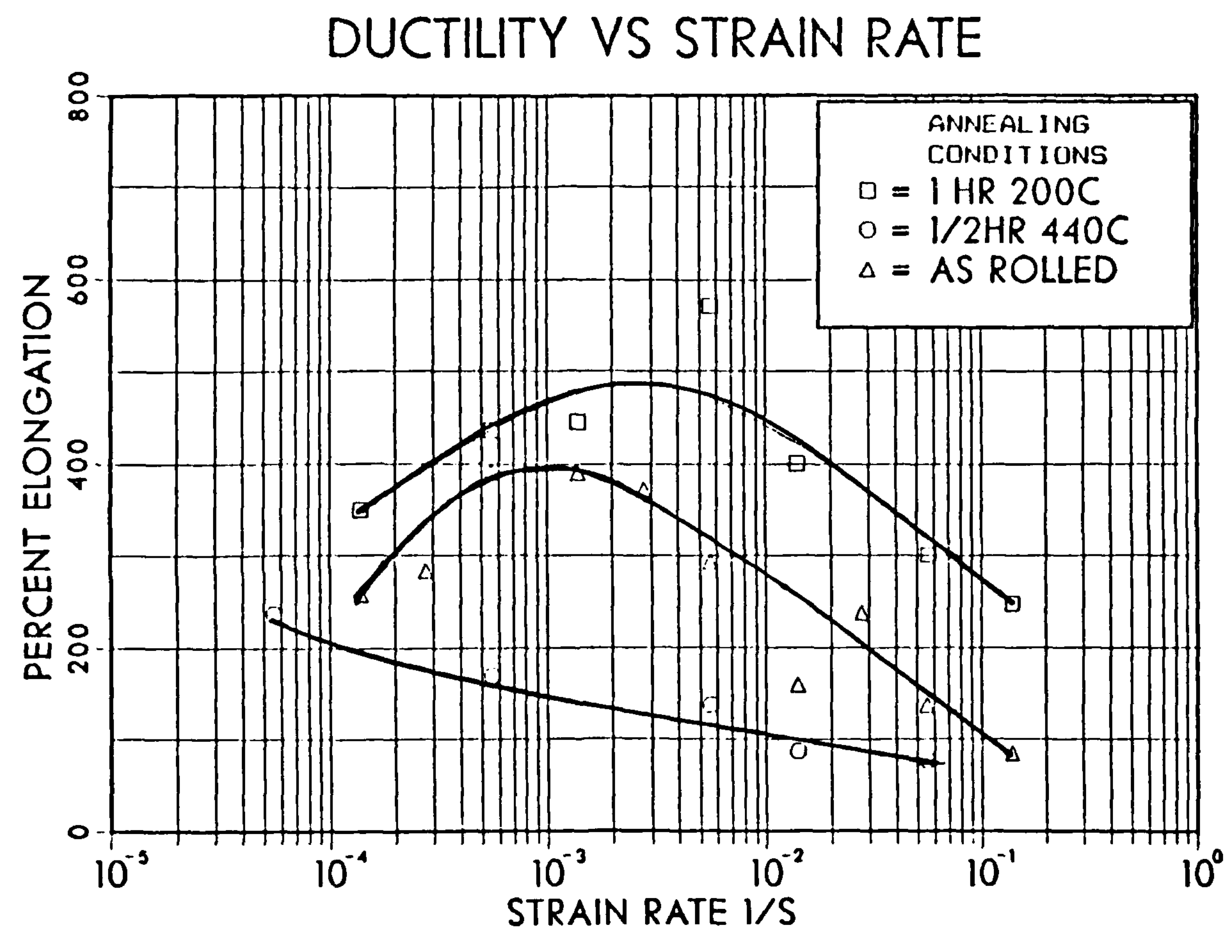
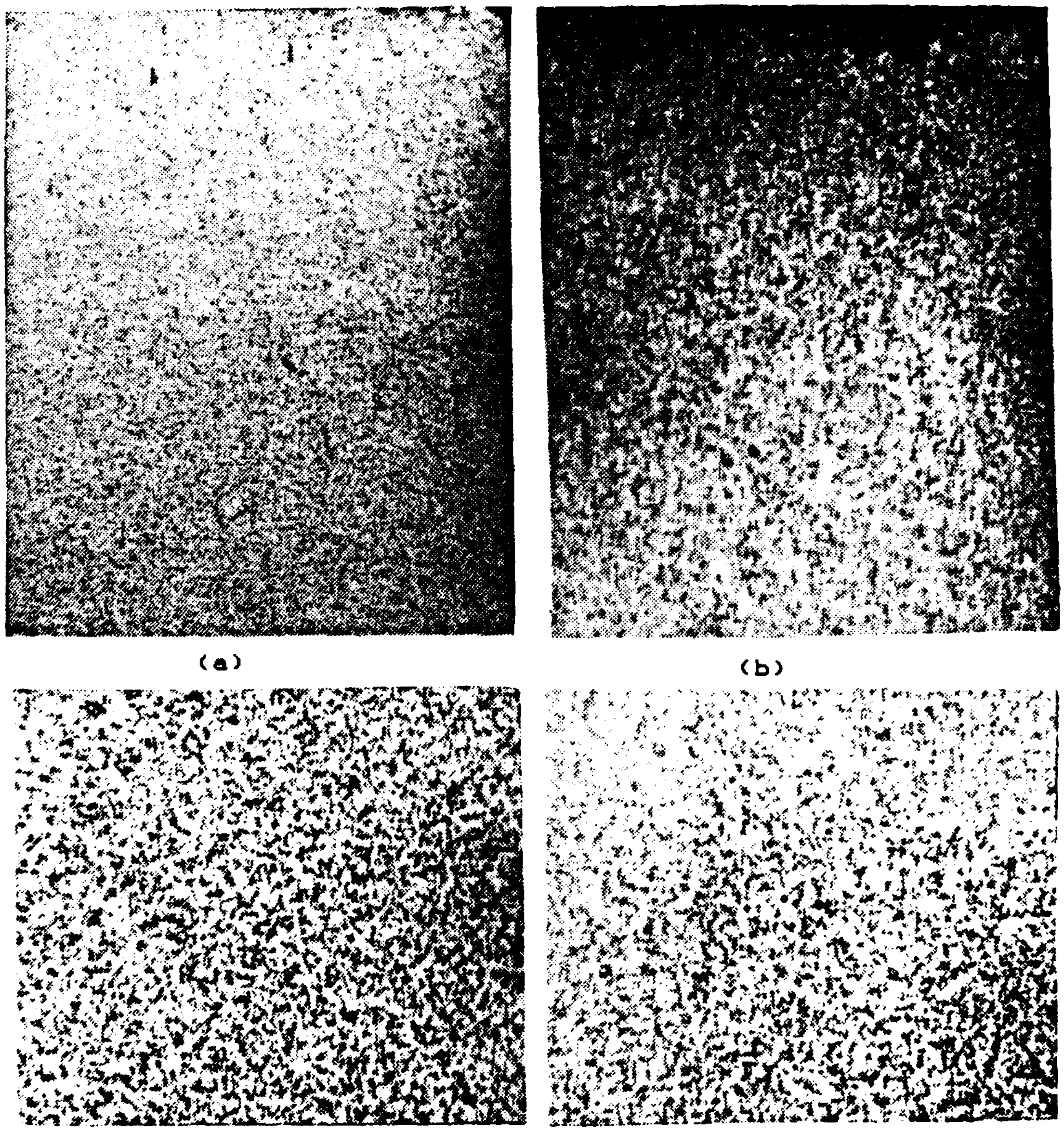


Figure 4.13 Ductility vs Strain Rate for Tensile Tests on Al-10.2%Mg-0.52%Mn Conducted at 300C. Comparing Data to Figure 4.12: Material Recrystallized by Annealing 1/2 Hour at 440C is Stronger and of Lower Ductility than As-Rolled Material at 300C. Annealing As-Rolled Material Below the Rolling Temperature Weakens the Material and Enhances the Superplastic Elongations.

ductility noted in Figure 4.14 would suggest this as a possibility. The principal microstructural effect expected from such are annealing treatment would be recovery of the subgrain structure. As noted previously, the dominant mechanism for 300 °C annealing was recovery. Also, annealing at 200 °C would be expected to produce further precipitation of the ß as the solubility of Mg decreases substantially in this temperature range (from 7.2% at 300 °C to about 3.5% at 200 °C). As such, it is expected that annealing below the rolling temperature would produce a finer recovered structure than would be attained by annealing at 300 °C. Further, the additional precipitation occurring at the lower temperature would result in further stabilization of the structure and hence a lesser tendency to coarsen during plastic flow. This still does not address the question of the mechanism responsible for the superplastic behavior of the rolled plus annealed material. Either both recrystallize during plastic deformation, with the result being an ultra fine grain size capable of deforming superplastically at 300 °C, or an alternate mechanism of superplastic flow must be invoked.

Figure 4.14 shows the microstructures of as-rolled material subsequently annealed for 1 hour at 200 °C and then tested at 300 °C at strain rates of 5.56x10-4 s-1 or $5.56\text{x}10^{-2}$ s^{-1}. Note very little cavitation is evident and the microstructure looks like that of the non-superplastic,



(a) (b) (c) (d)

Figure 4.14 As-Rolled Material that has been Annealed for 1 Hour at 200 °C and Tested at 300 °C at Strain Rates of 5.56×10^{-4} (a) & (c), and 5.56×10^{-2} (b) & (d). Note Very Little Cavitation is Evident and Microstructure Looks Like the Non-Superplastic Recrystallized Material Tested at 300 °C (Figure 4.8). However, Unlike the Material in Figure 4.8 This Material is Very Superplastic. Elongation of 437% for (a) & (c) and 401% for (b) & (d). Magnification for (a) & (b) is x64 and for (c) & (d) is x250, Graf-Sargent Etch.

recrystallized material also tested at 300 °C (Figure 4.8). However, unlike the recrystallized material in Figure 4.8, this material is very superplastic. An elongation of 572% was observed at a strain rate of $5.56 \times 10^{-3}\ s^{-1}$, and of special note is the absence of any cavitation in the structure. Additional data on other annealing treatments in this research are contained in Appendix A.

In summary, this research has found that the recrystallized Al-10.2%Mg-0.52%Mn alloy behaves in much the same manner as other aluminum-base alloys using a thermo-mechanical processing treatment similar to current practice i.e. warm rolling, recrystallization above the solvus and then deforming above the solvus temperature. However, the as-rolled material tested at 300 °C or as-rolled material which has been annealed for 1 hour at 200 °C prior to testing at 300 °C, is very superplastic where as recrystallized material tested at 300 °C is not superplastic at all.

V. CONCLUSIONS AND RECOMMENDATIONS

Two significant conclusions can be drawn from this research: 1) Strain rate sensitivity and the ductility of the as-rolled material can be controlled and improved by annealing below the rolling temperature (i.e. in this research for 1 hour at 200 °C). Annealing below the warm rolling temperature may allow as much as an additional 4.0 wt. pct. Mg to precipitate out of solution in the form of fine Al_8Mg_5 (β) which would tend to stabilize grain/subgrain boundaries during subsequent tension testing at 300 °C. 2) The strength and ductility of warm rolled and recrystallized material follows a pattern as a function of strain rate and temperature very similar to that observed in other superplastic Al-alloys in that extensive superplasticity is observed only at relatively high temperatures (above the solvus for the strengthening component); in contrast, the warm rolled condition exhibits superplasticity at much lower temperatures and shows little or no tendency to cavitate during superplastic flow. Under identical conditions the recrystallized material is not superplastic.

Continued research in the following areas is indicated: 1) transmission electron microscopy is required for examination of the grain/subgrain boundaries and to

examine the effect of deformation at elevated temperature on these structures; 2) annealing time-temperature variables following warm rolling need further investigation to optimize superplastic properties; 3) applications of the thermomechanical processing used in this thesis to current processing of high strength aluminum alloys such as 7075 or 7475, which are recrystallized before forming.

APPENDIX A

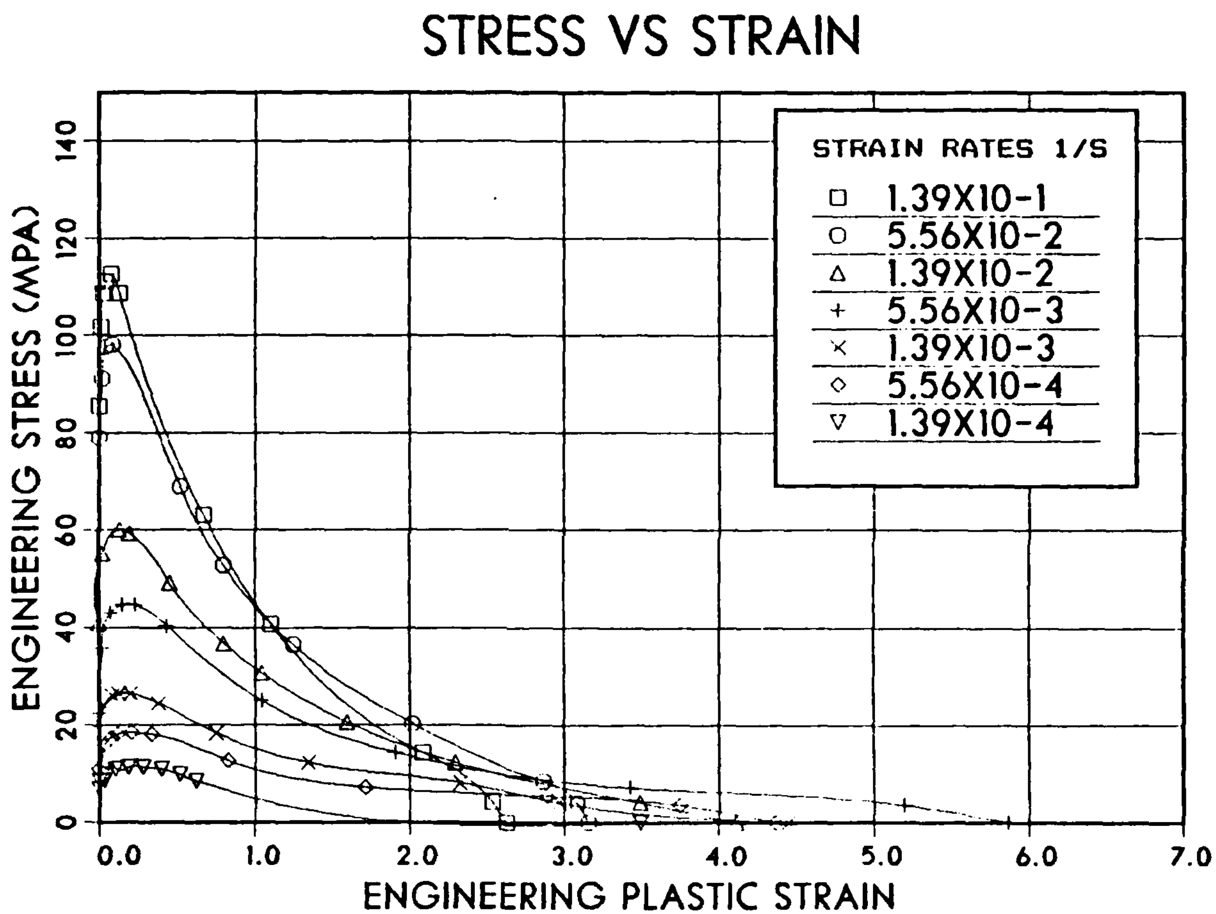


Figure A.1 Engineering Stress vs Engineering Plastic Strain for Tensile Tests Conducted at 300C on Al-10.2%Mg-0.52%Mn. The Material was Warm Rolled at 300C and then Annealed for 1 Hour at 200C Prior to Tension Testing at 300C.

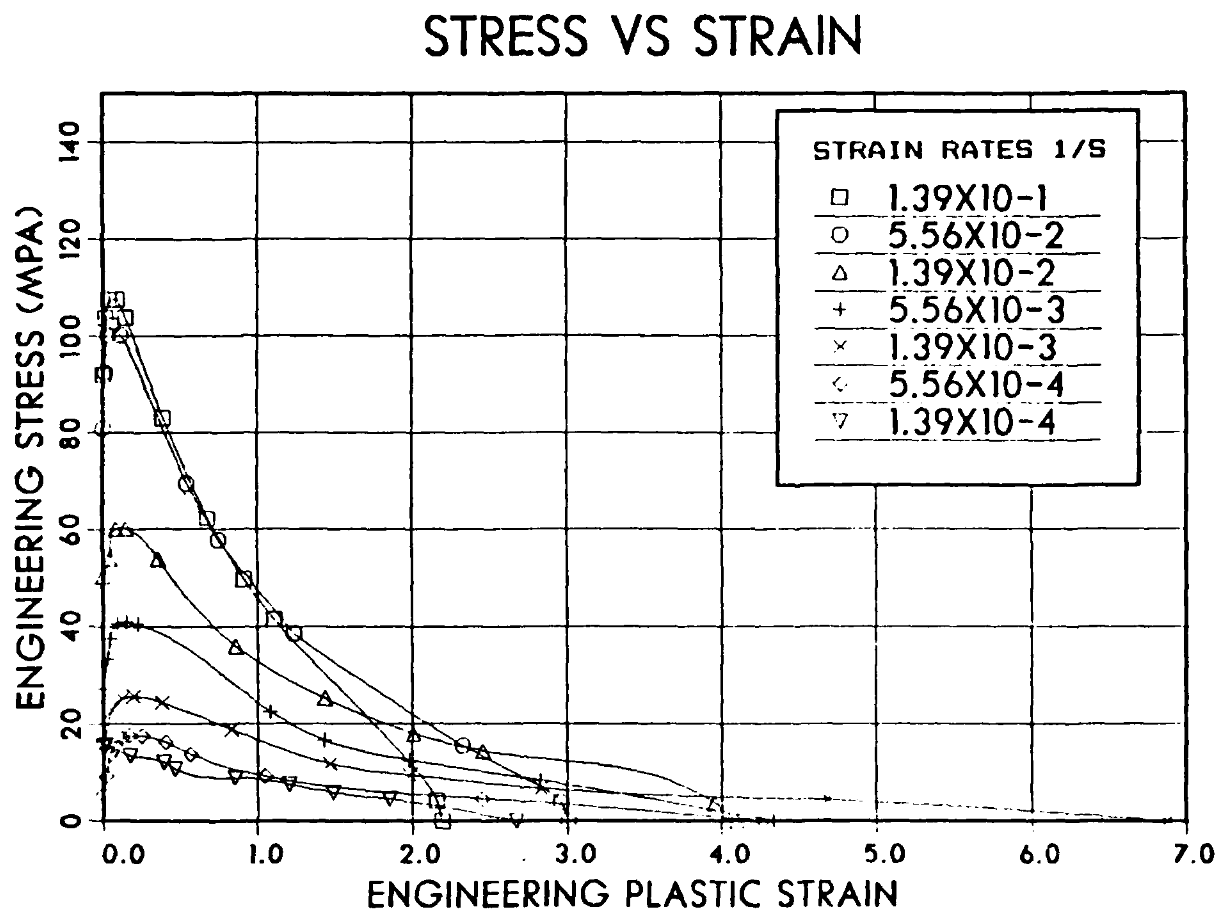


Figure A.2 Engineering Stress vs Engineering Plastic Strain for Tensile Tests Conducted at 300C on Al-10.2%Mg-0.52%Mn. The Material was Warm Rolled at 300C and then Annealed for 10 Hours Prior to Tension Testing at 300C.

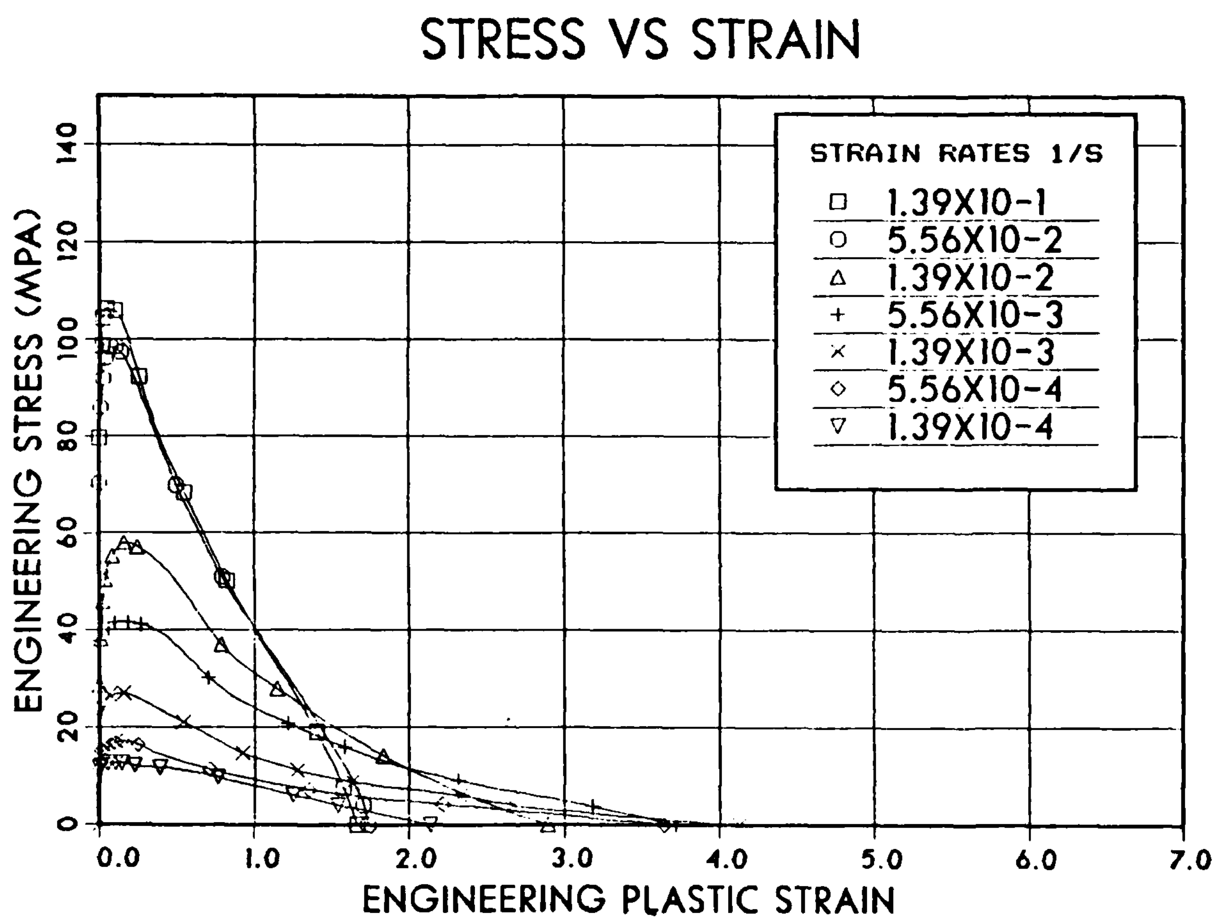


Figure A.3 Engineering Stress vs Engineering Plastic Strain for Tensile Tests Conducted at 300C on Al-10.2%Mg-0.52%Mn. The Material was Warm Rolled at 300C and then Annealed for 1/2 Hour at 250C Prior to Tension Testing at 300C.

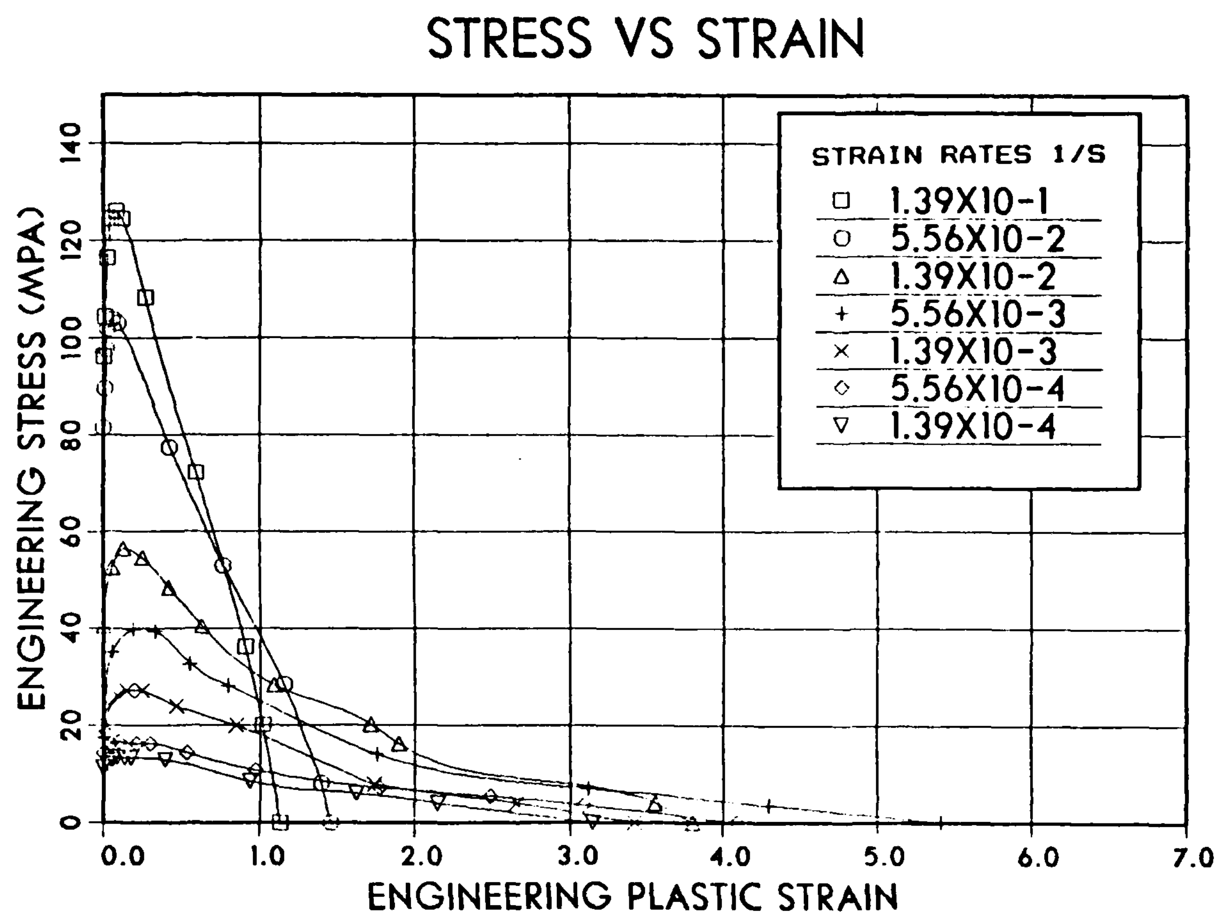


Figure A.4 Engineering Stress vs Engineering Plastic Strain for Tensile Tests Conducted at 300C on Al-10.2%Mg-0.52%Mn. The Material was Warm Rolled at 300C and then Annealed for 1 Hour at 250C Prior to Tension Testing at 300C.

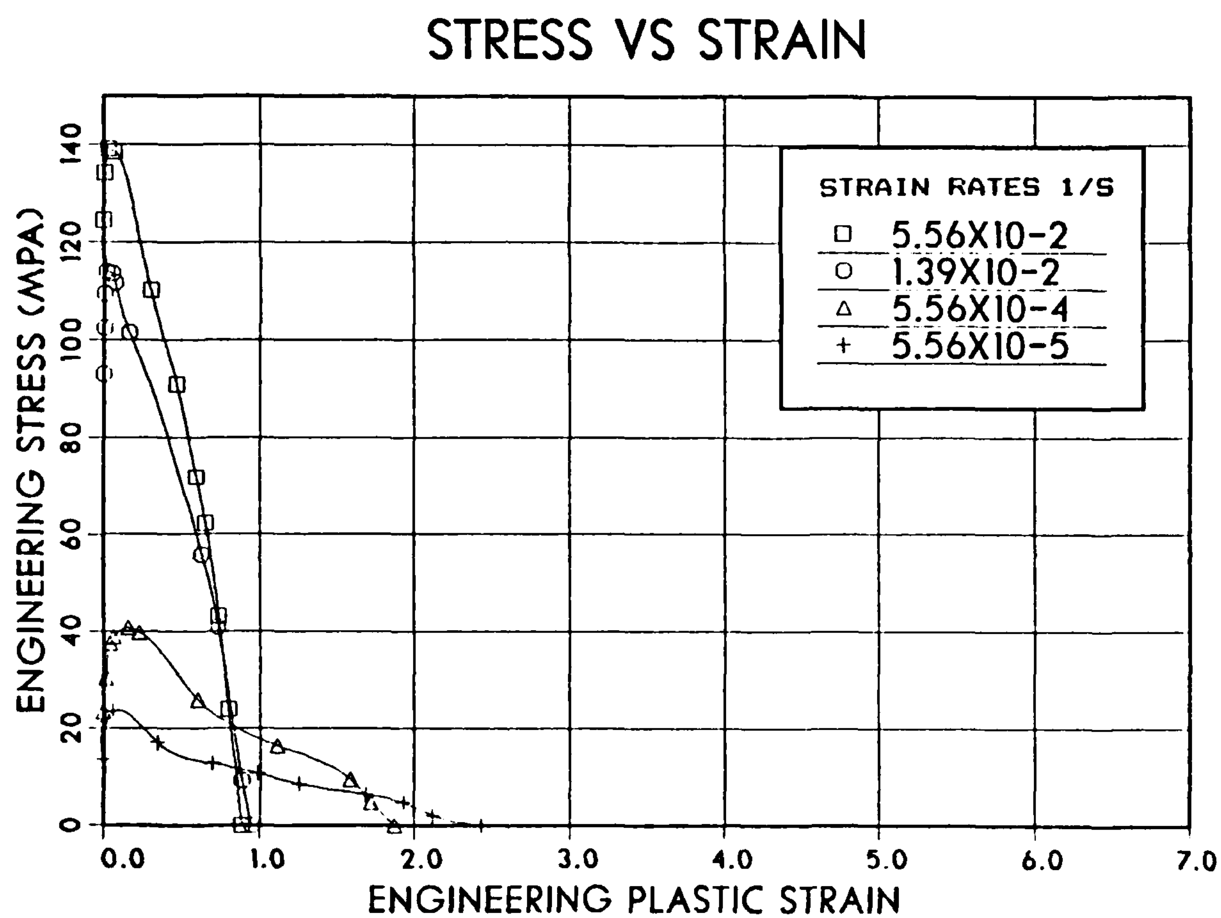


Figure A.5 Engineering Stress vs Engineering Plastic Strain for Tensile Tests Conducted at 300C on Al-10.2%Mg-0.52%Mn. The Material was Warm Rolled at 300C and then Annealed for 1/2 at 440C Prior to Tension Testing at 300C.

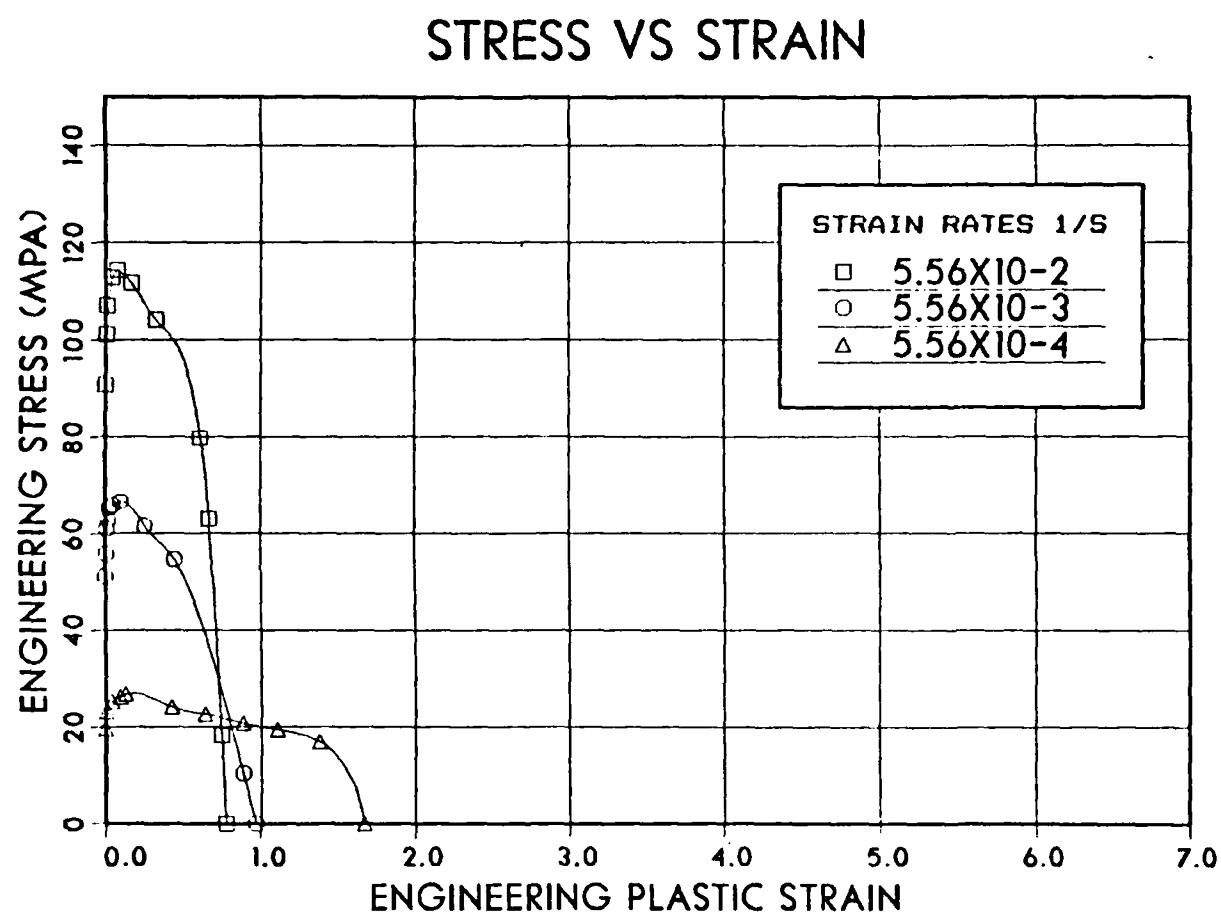


Figure A.6 Engineering Stress vs Engineering Plastic Strain for Tensile Tests Conducted at 325C on Al-10.2%Mg-0.52%Mn. The Material was Warm Rolled at 300C and then Annealed for 1/2 Hour at 440C Prior to Tension Testing at 325C.

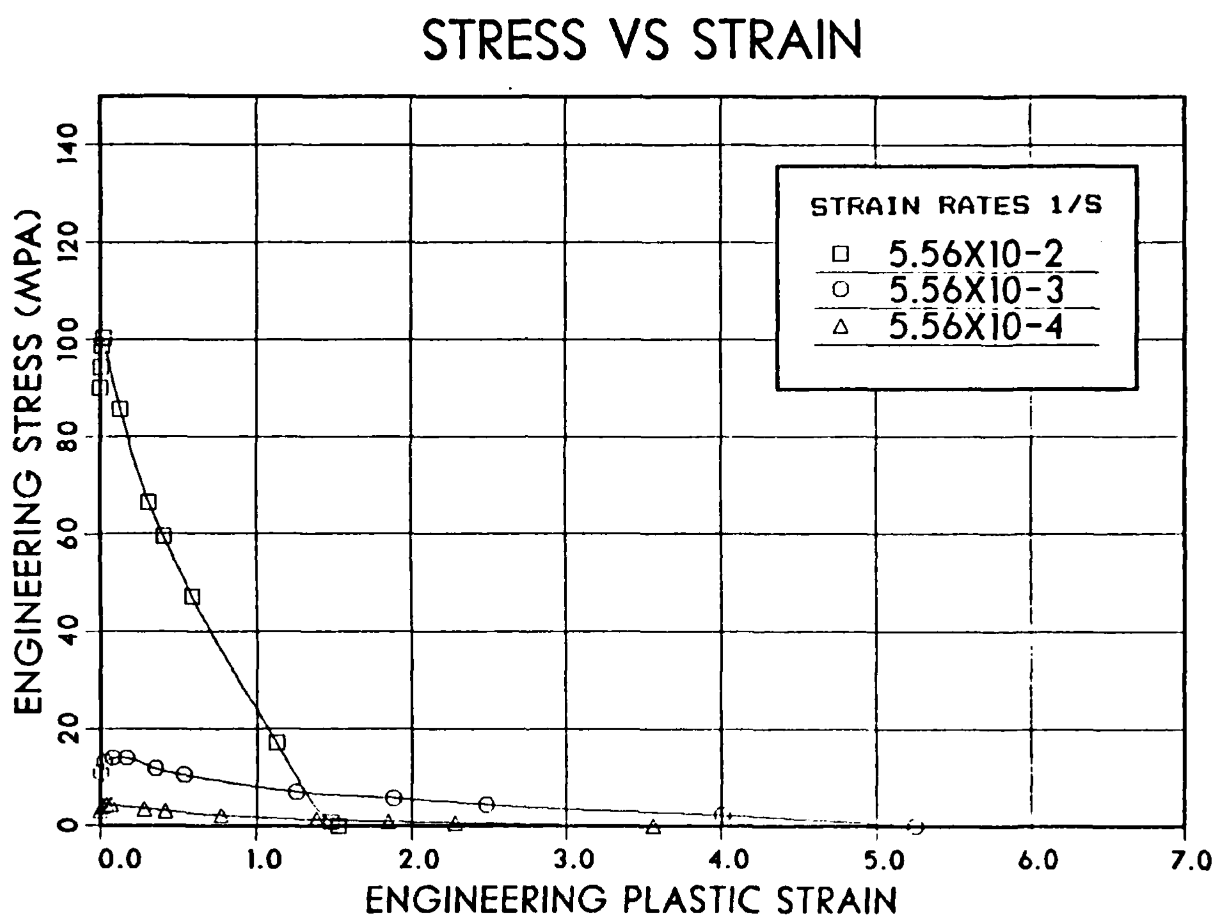


Figure A.7 Engineering Stress vs Engineering Plastic Strain for Tensile Tests Conducted at 350C on Al-10.2%Mg-0.52%Mn. The Material was Warm Rolled at 300C and then Annealed for 1/2 Hour at 440C Prior to Tension Testing at 350C.

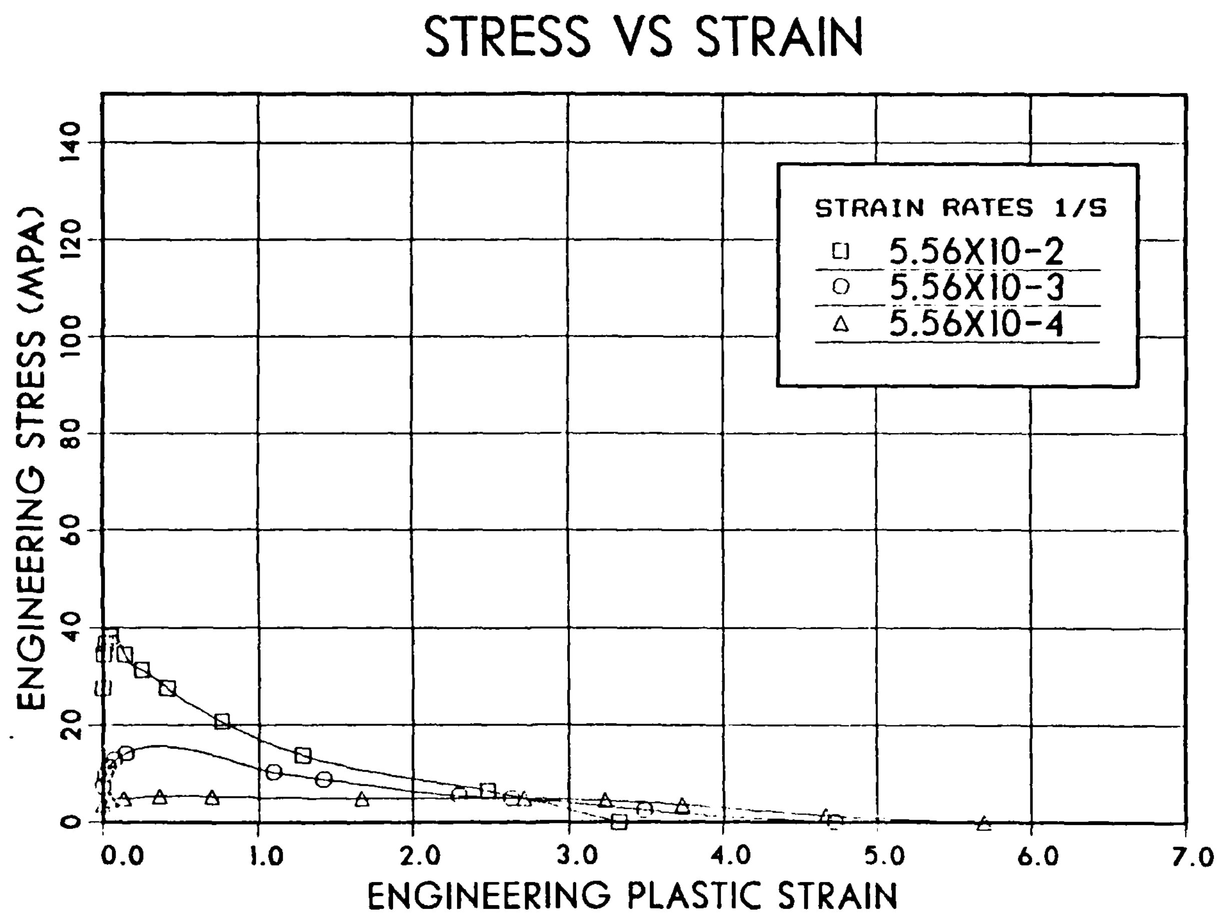


Figure A.8 Engineering Stress vs Engineering Plastic Strain for Tensile Tests Conducted at 425C on Al-10.2%Mg-0.52%Mn. The Material was Warm Rolled at 300C and then Annealed for 1/2 Hour at 440C Prior to Tension Testing at 425C.

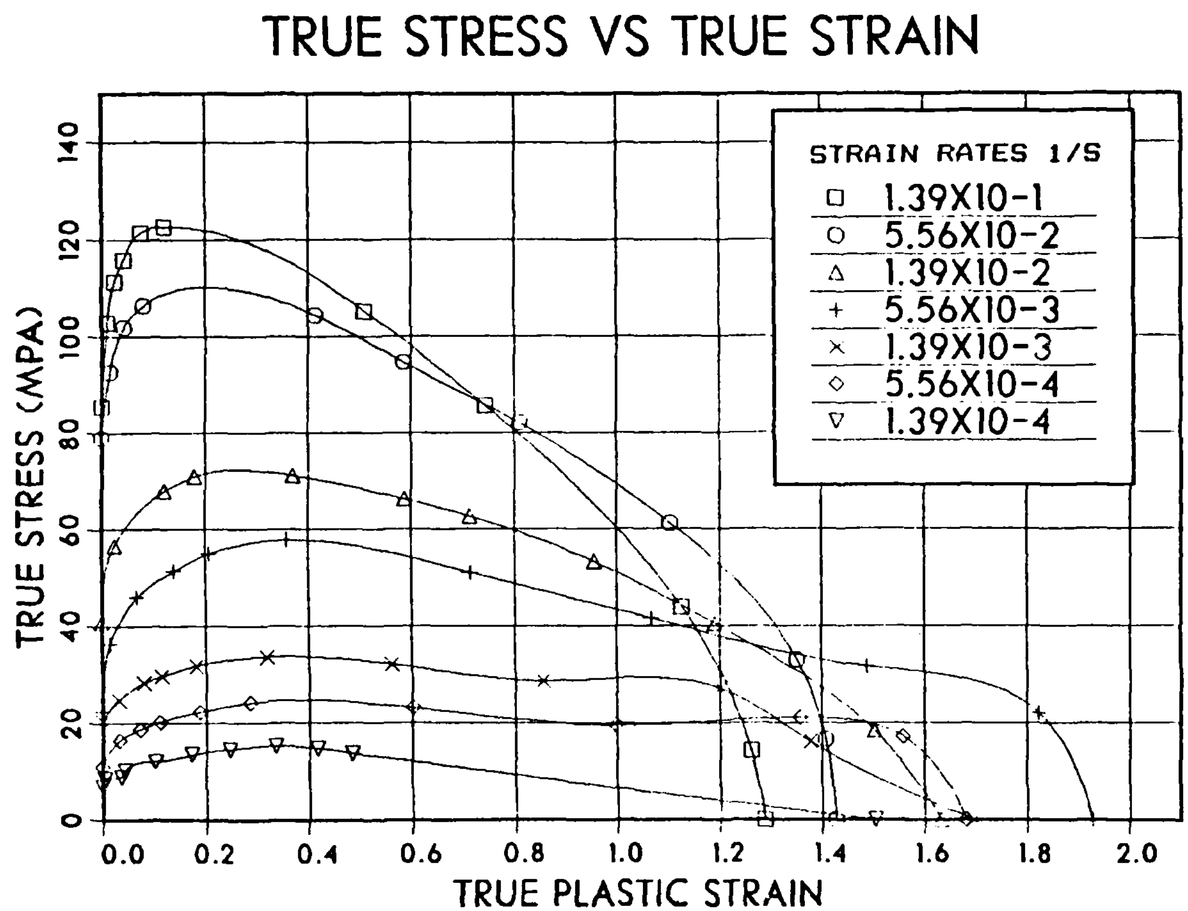


Figure A.9 True Stress vs True Plastic Strain for Tensile Tests Conducted at 300C on Al-10.2%Mg-0.52%Mn. The Material was Warm Rolled at 300C and then Annealed for 1 Hour at 200C Prior to Tension Testing at 300C.

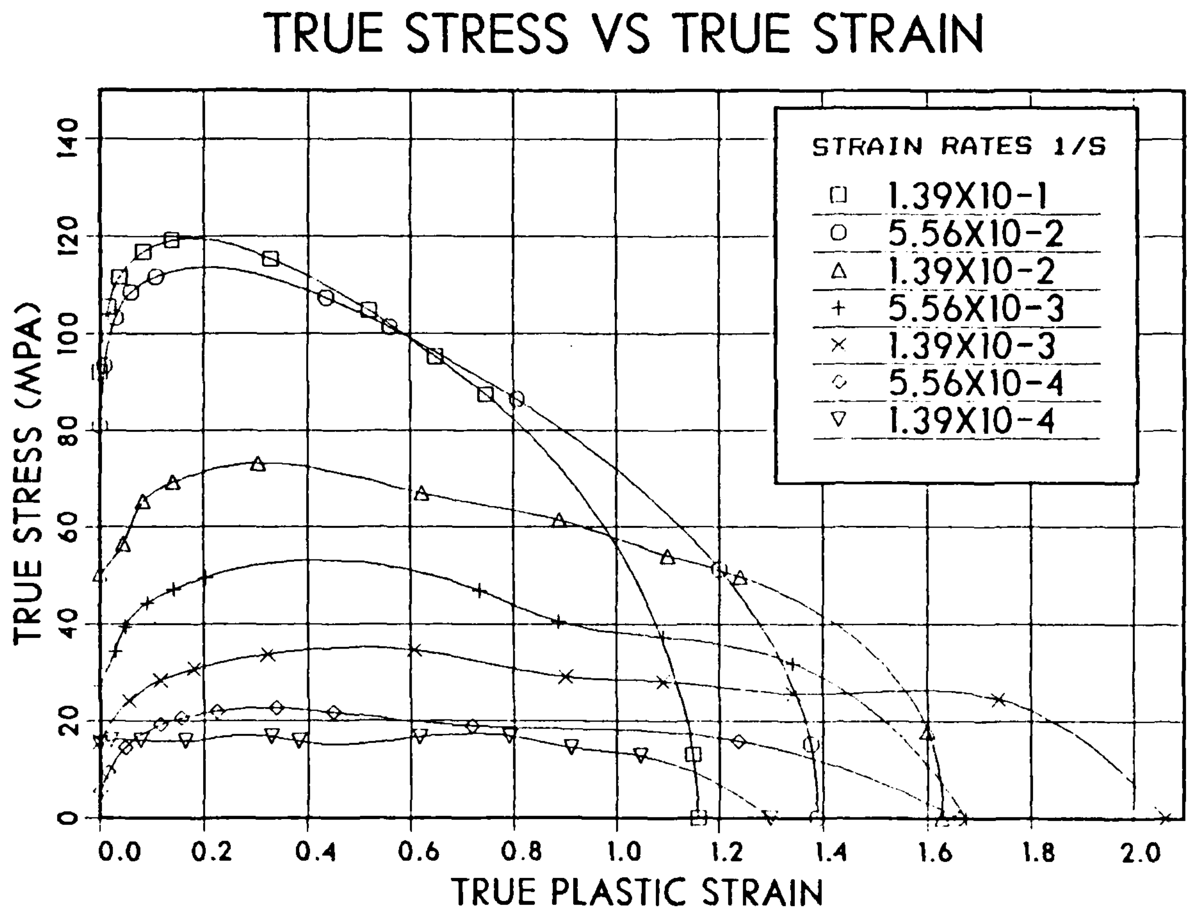


Figure A.10 True Stress vs True Plastic Strain for Tensile Tests Conducted at 300C on Al-10.2%Mg-0.52%Mn. The Material was Warm Rolled at 300C and then Annealed for 10 Hours at 200C Prior to Tension Testing at 300C.

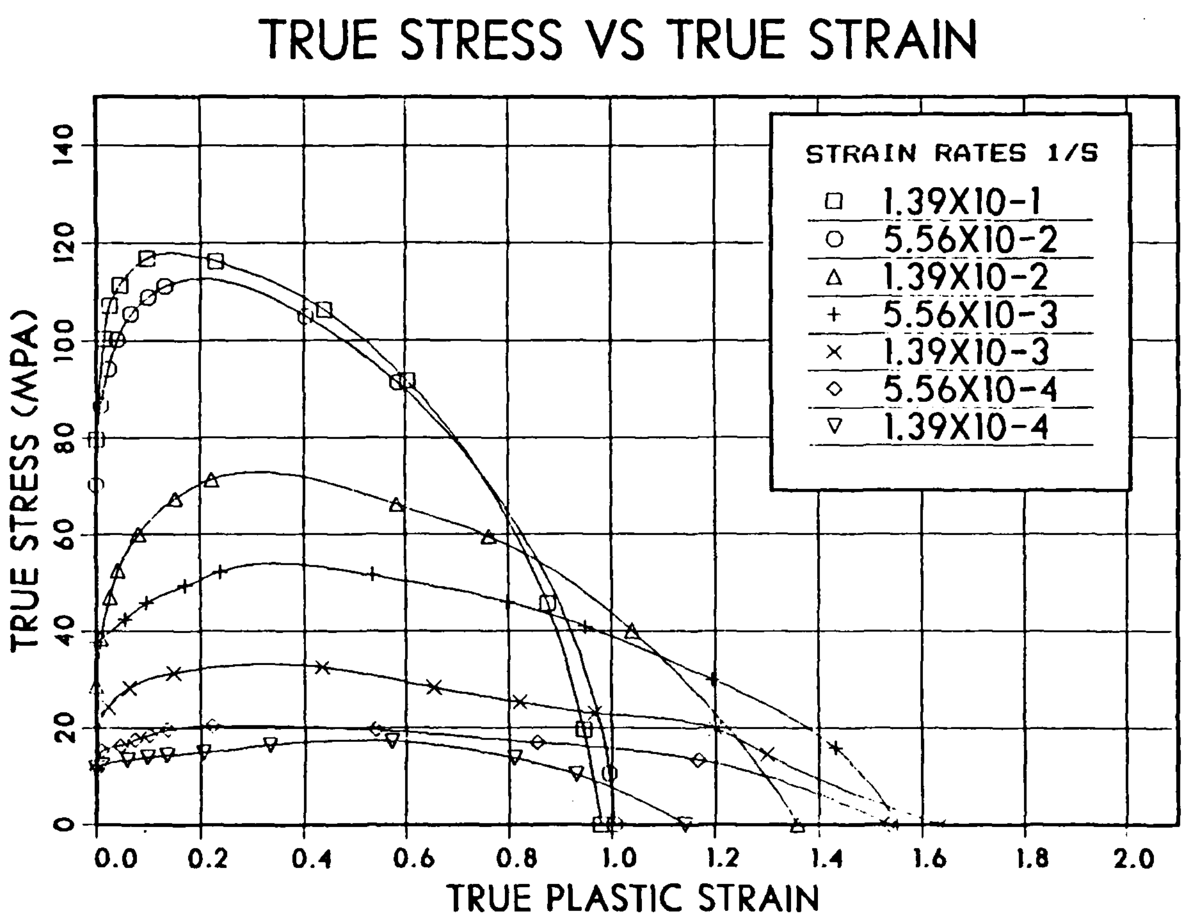


Figure A.11 True Stress vs True Plastic Strain for Tensile Tests Conducted at 300C on Al-10.2%Mg-0.52%Mn. The Material was Warm Rolled at 300C and then Annealed for 1/2 Hour at 250C Prior to Tension Testing at 300C.

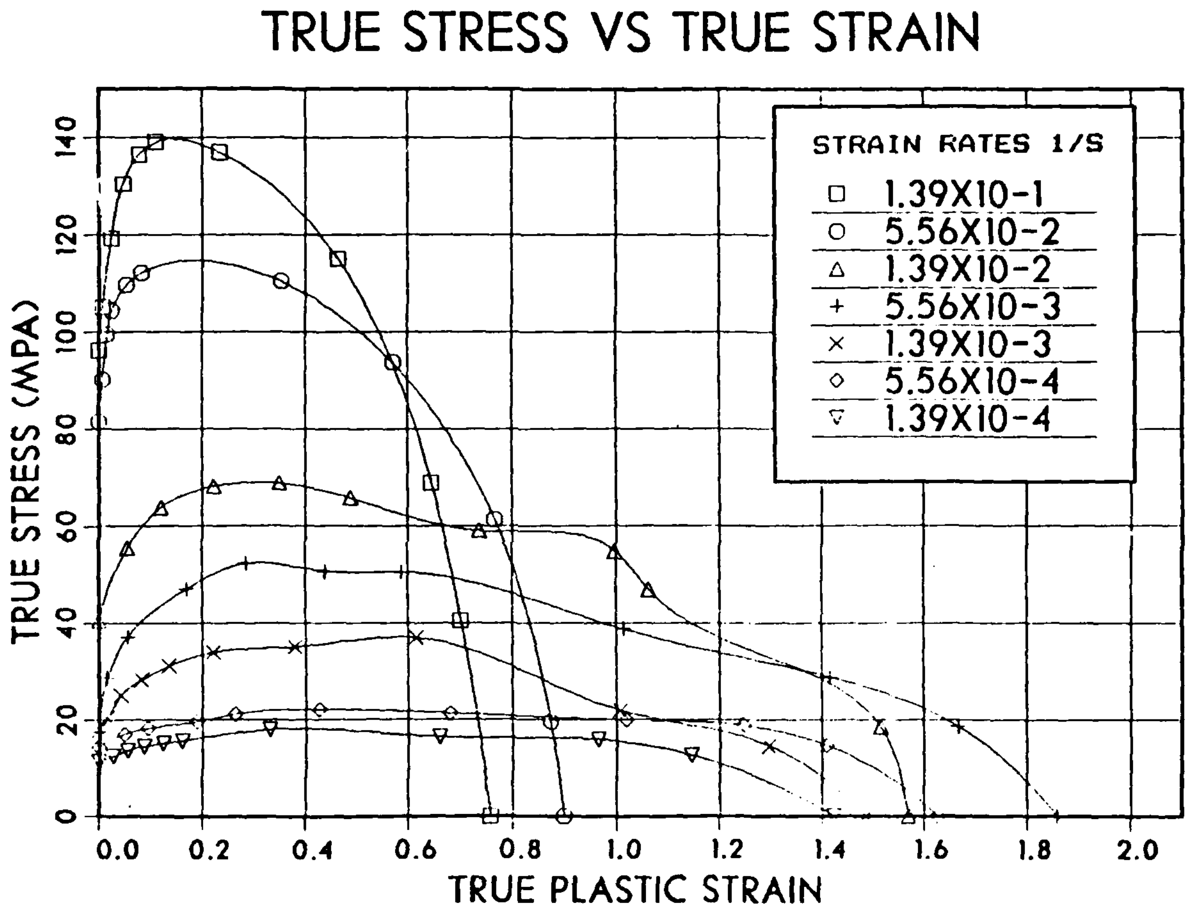


Figure A.12 True Stress vs True Plastic Strain for Tensile Tests Conducted at 300C on Al-10.2%Mg-0.52%Mn. The Material was Warm Rolled at 300C and then Annealed for 1 Hour at 250C Prior to Tension Testing at 300C.

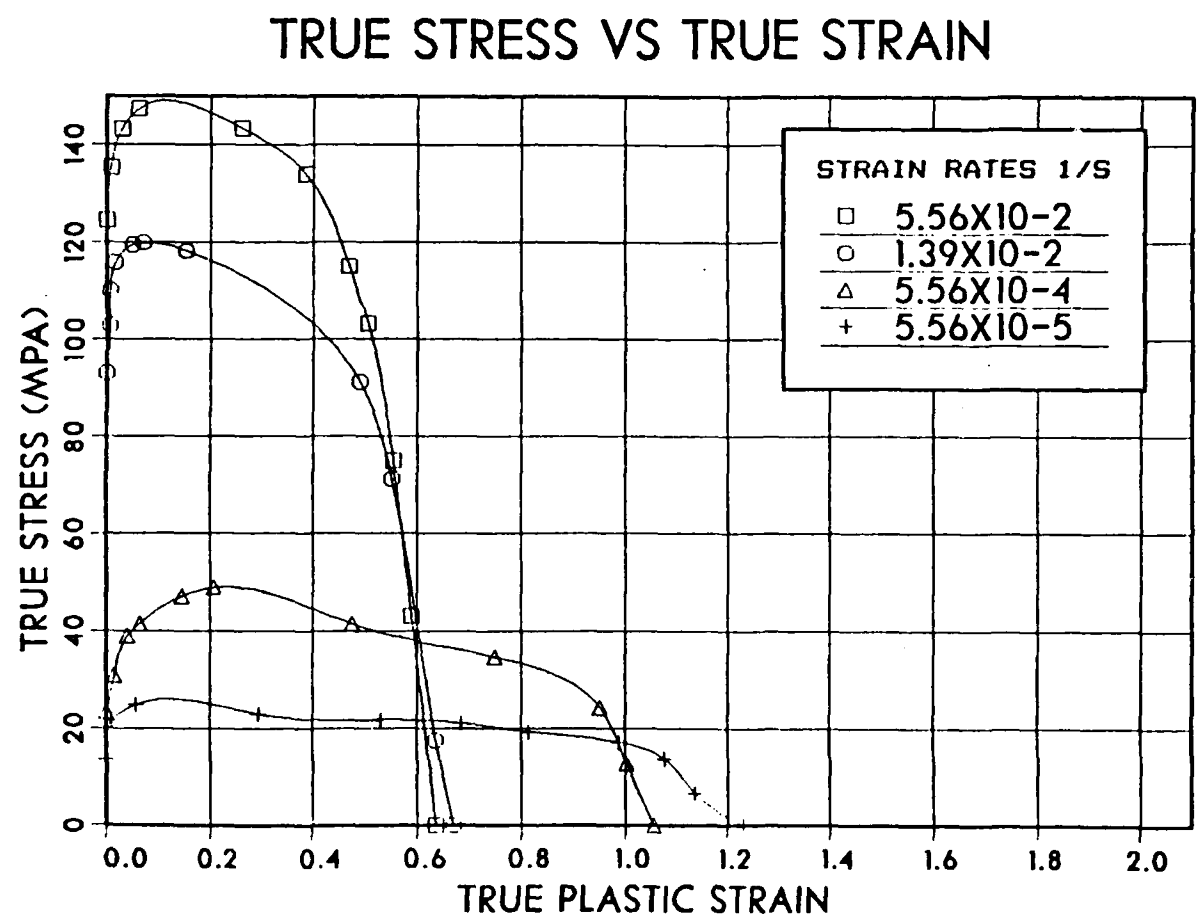


Figure A.13 True Stress vs True Plastic Strain for Tensile Tests Conducted at 300C on Al-10.2%Mg-0.52%Mn. The Material was Warm Rolled at 300C and then Annealed for 1/2 Hour at 440C Prior to Tension Testing at 300C.

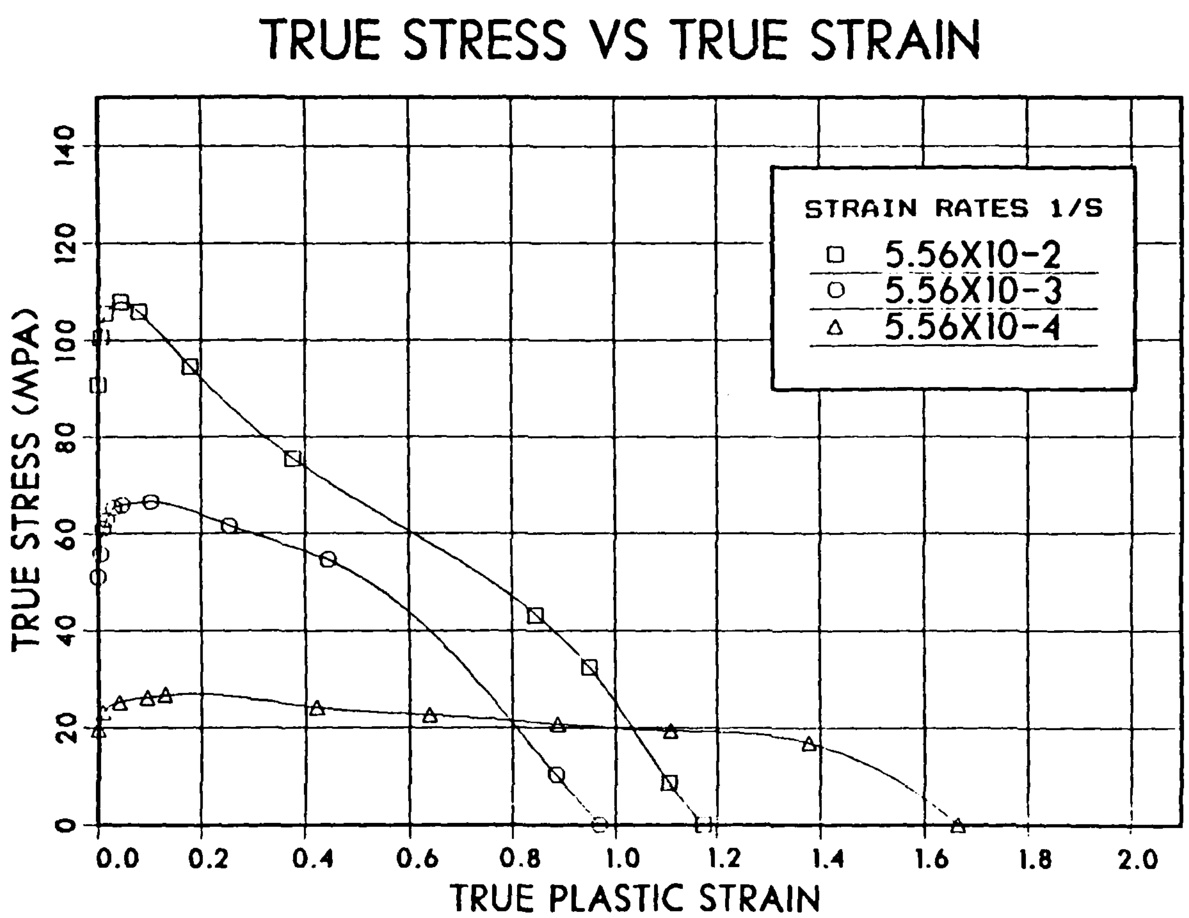


Figure A.14 True Stress vs True Plastic Strain for Tensile Tests Conducted at 325C on Al-10.2%Mg-0.52%Mn. The Material was Warm Rolled at 300C and then Annealed for 1/2 Hour at 440C Prior to Tension Testing at 325C.

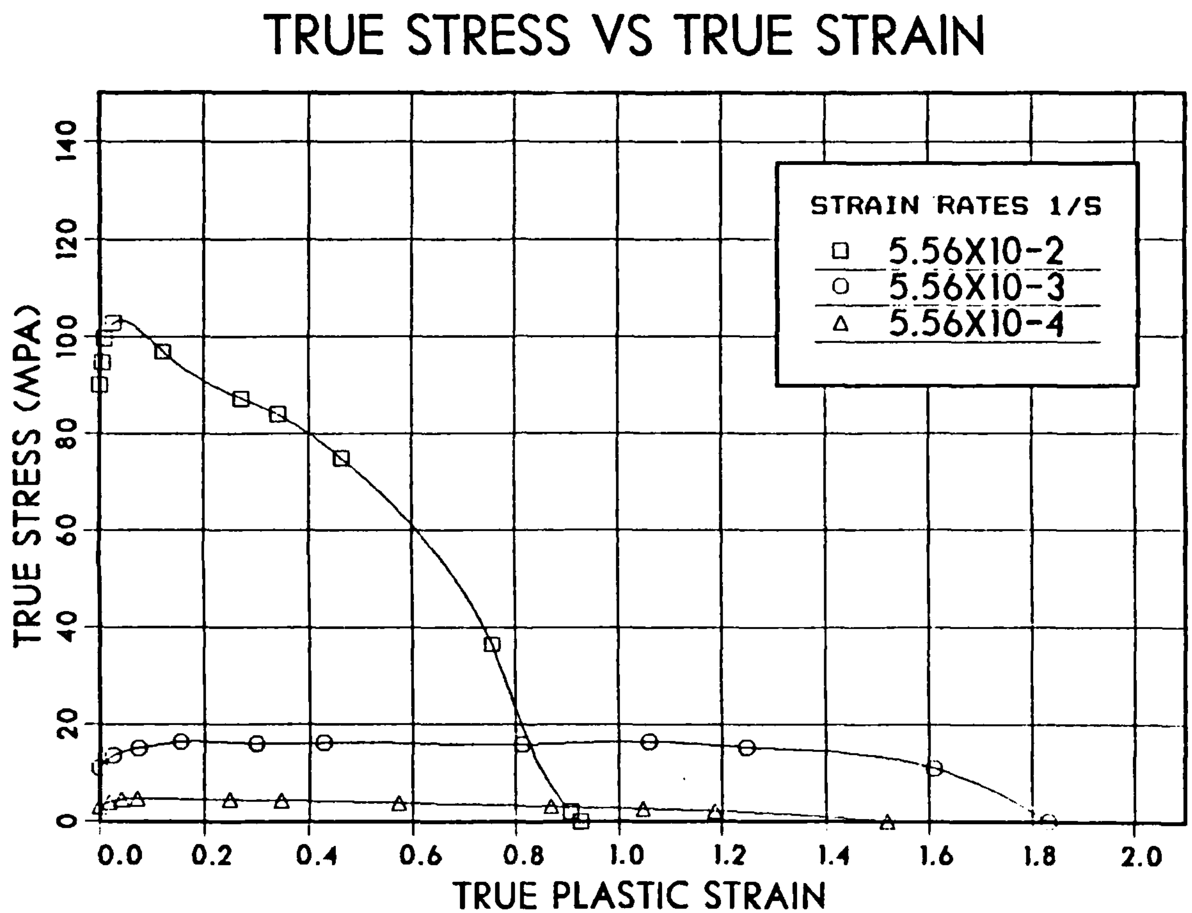


Figure A.15 True Stress vs True Plastic Strain for Tensile Tests Conducted at 350C on Al-10.2%Mg-0.52%Mn. The Material was Warm Rolled at 300C and then Annealed for 1/2 Hour at 440C Prior to Tension Testing at 350C.

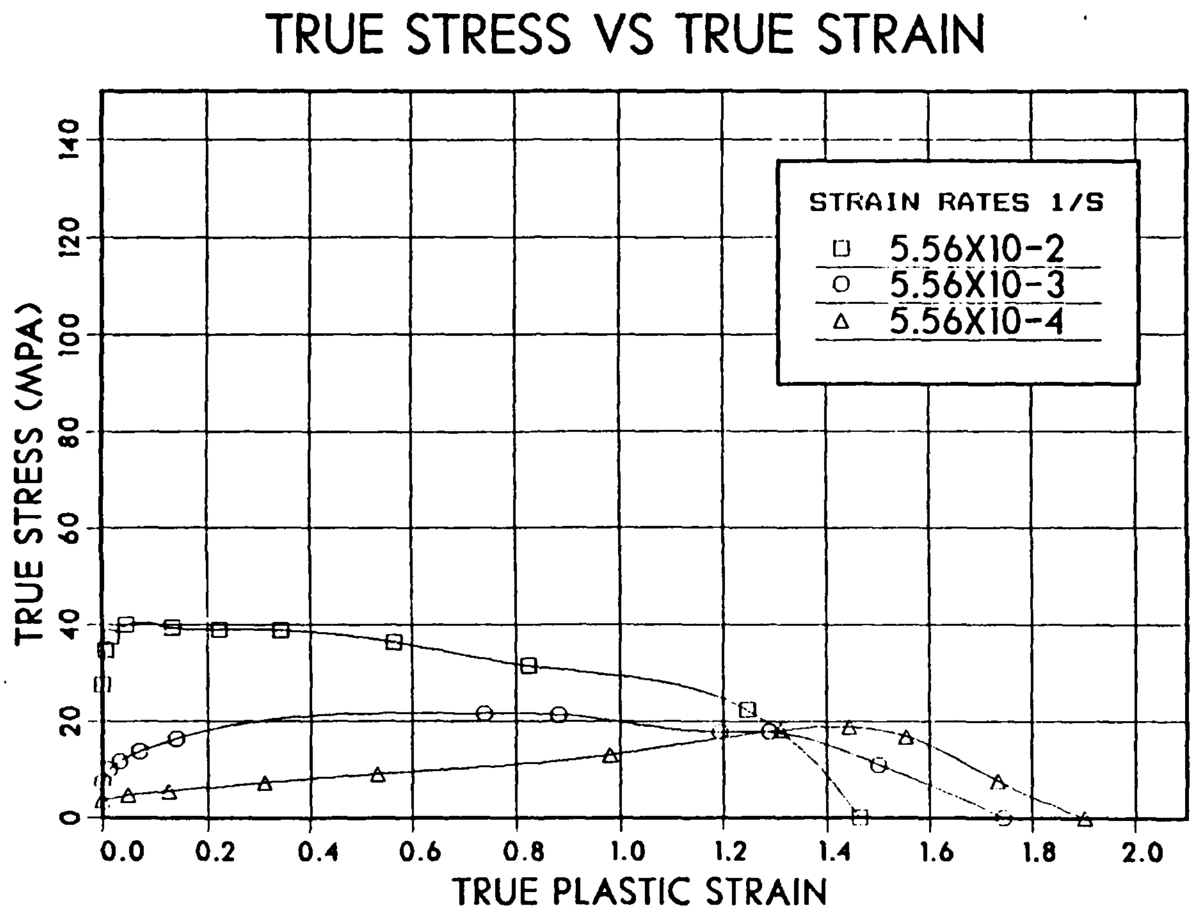


Figure A.16 True Stress vs True Plastic Strain for Tensile Tests Conducted at 425C on Al-10.2%Mg-0.52%Mn. The Material was Warm Rolled at 300C and then Annealed for 1/2 Hour at 440C Prior to Tension Testing at 425C.

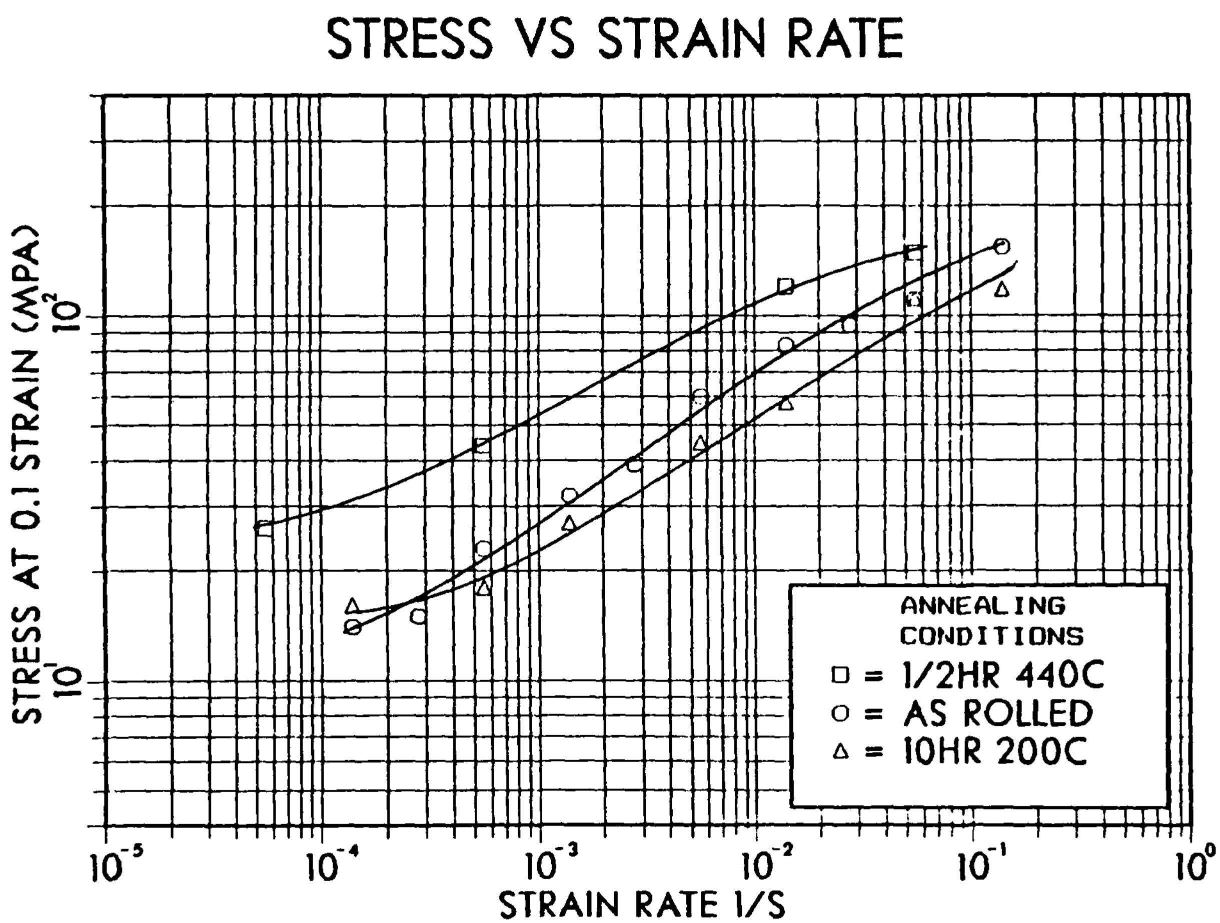


Figure A.17 True Stress at 0.1 Strain vs Strain Rate for Tensile Tests on Al-10.2%Mg-0.52% Mn Conducted at 300C. Annealing for 1/2 Hour at 440C Results in a Recrystallized Structure Which is Stronger than the As-Rolled Condition While Annealing for 10 Hours at 200C Results in Lower Strength when Compared to As-Rolled Material.

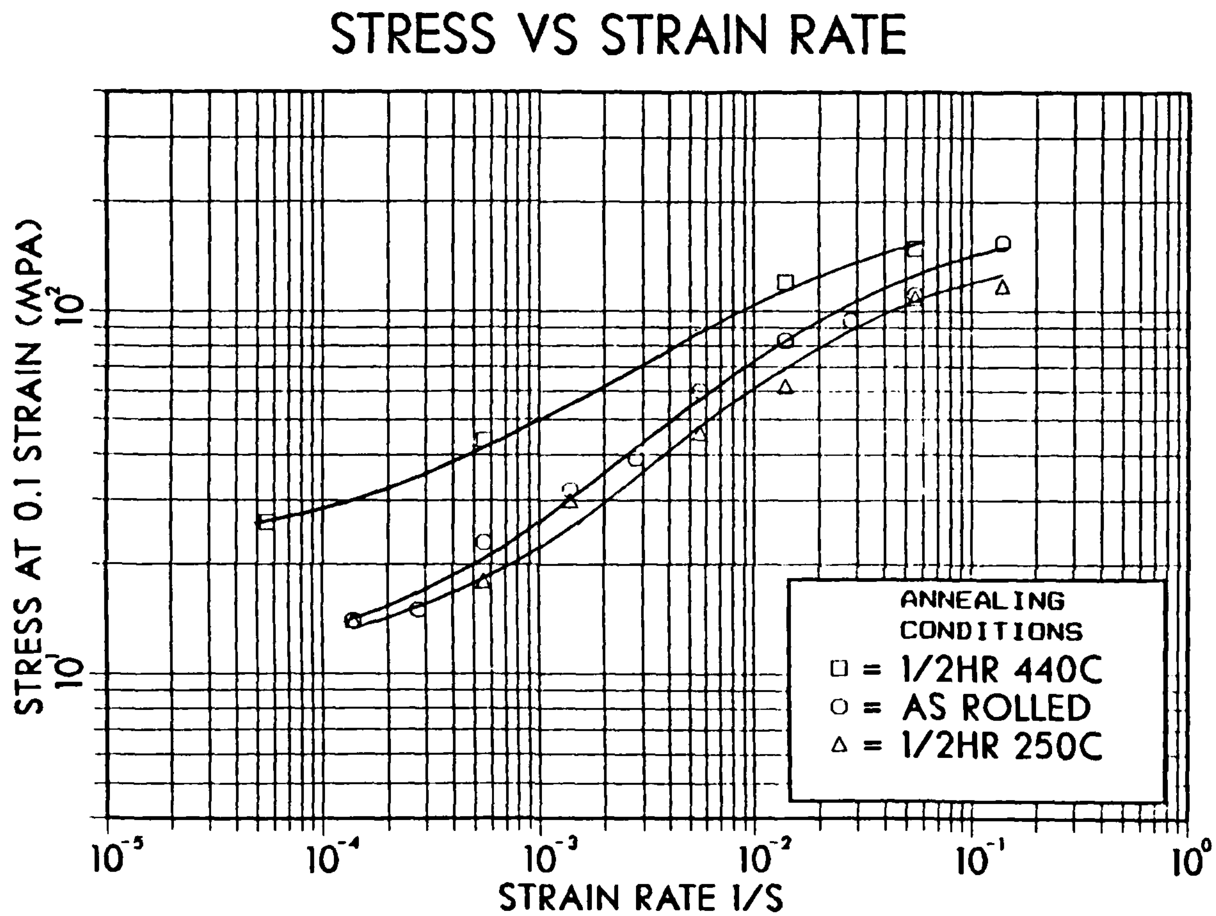


Figure A.18 True Stress at 0.1 Strain vs Strain Rate for Tensile Tests on Al-10.2%Mg-0.52%Mn Conducted at 300C. Annealing for 1/2 Hour at 440C Results in a Recrystallized Structure Which is Stronger than the As-Rolled Condition While Annealing for 1/2 Hour at 250C Results in Lower Strength when Compared to As-Rolled Material.

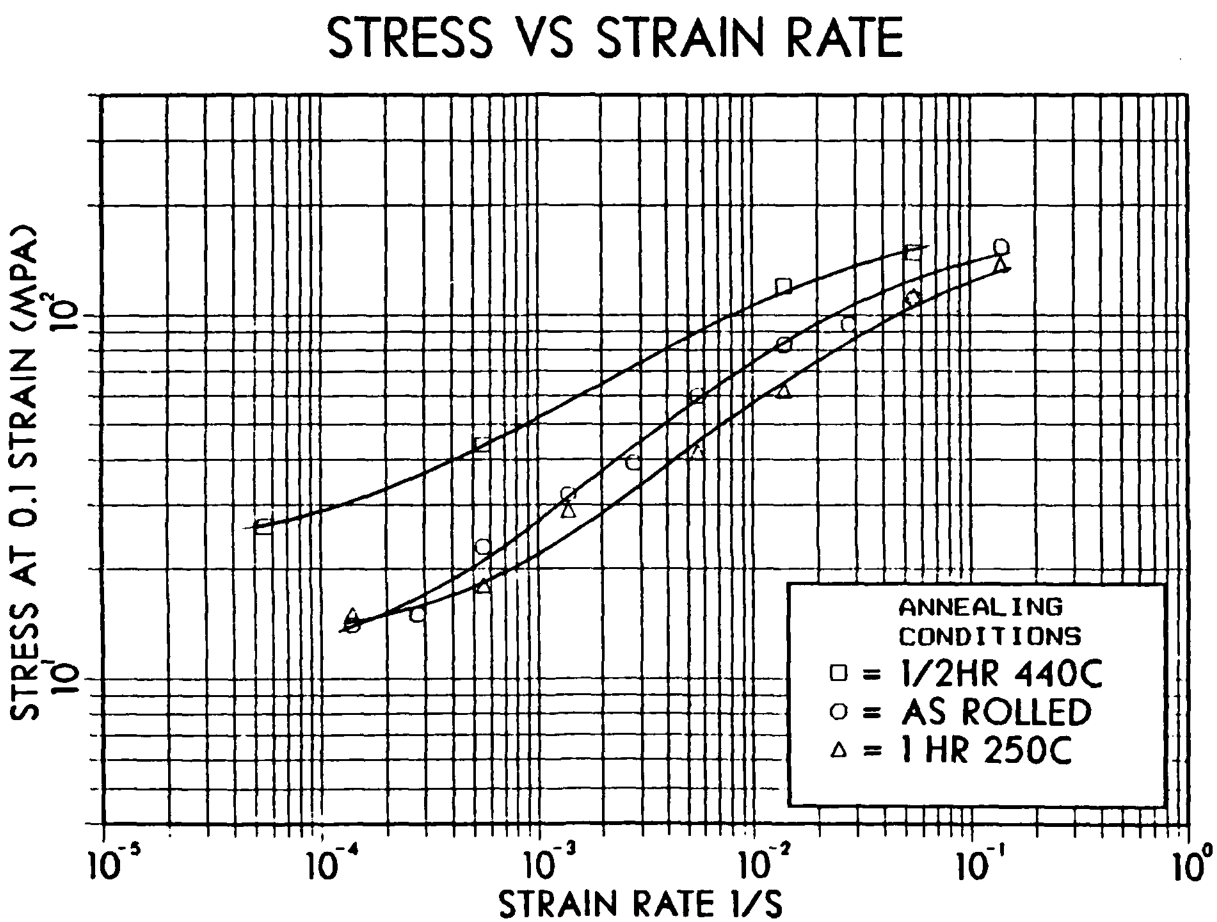


Figure A.19 True Stress at 0.1 Strain vs Strain Rate for Tensile Tests on Al-10.2%Mg-0.52%Mn Conducted at 300C. Annealing for 1/2 Hour at 440C Results in a Recrystallized Structure Which is Stronger than the As-Rolled Condition While Annealing for 1 Hour at 250C Results in Lower Strength when Compared to As-Rolled Material.

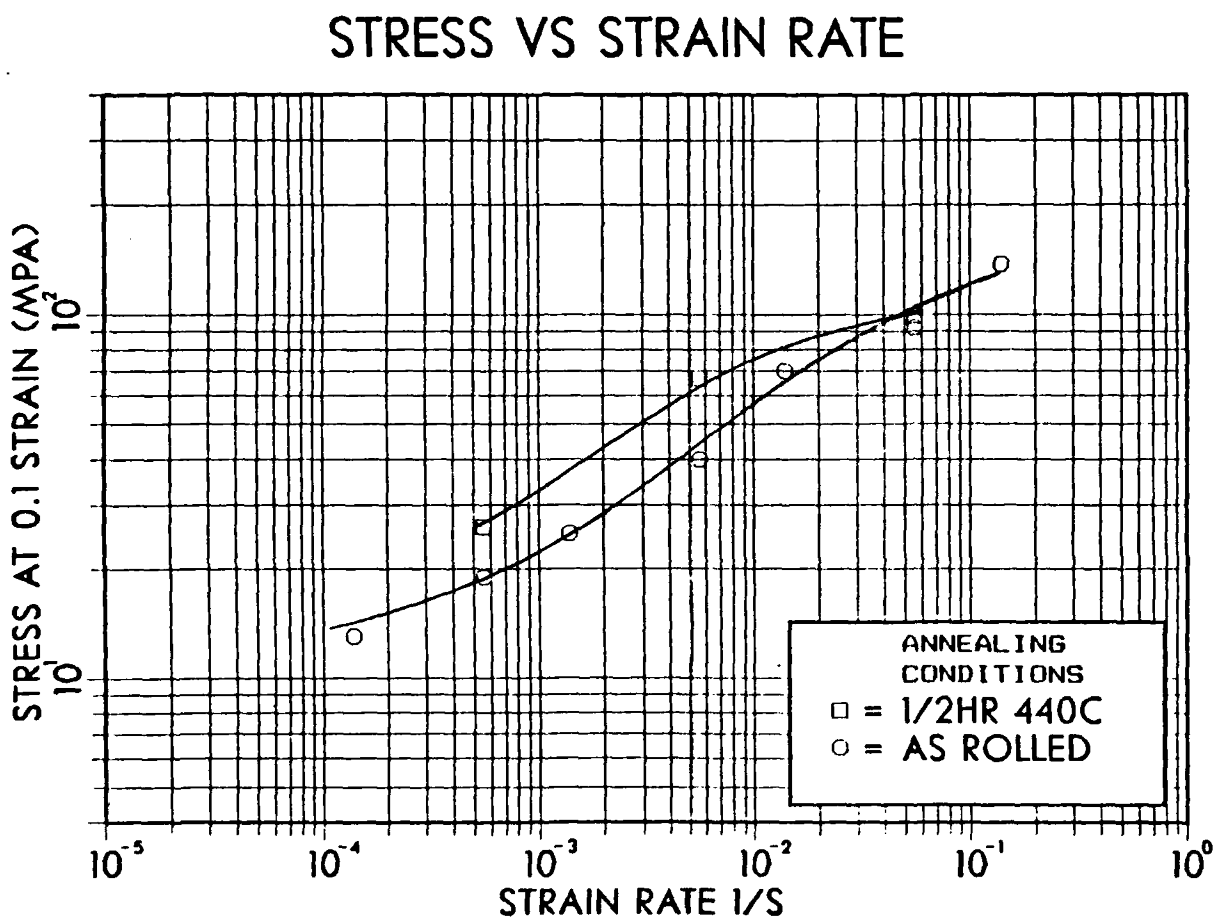


Figure A.20 True Stress at 0.1 Strain vs Strain Rate for Tensile Tests on Al-10.2%Mg-0.52%Mn Conducted at 325C. The As-Rolled Condition is Weaker than Material Recrystallized by Annealing 1/2 Hour at 440C.

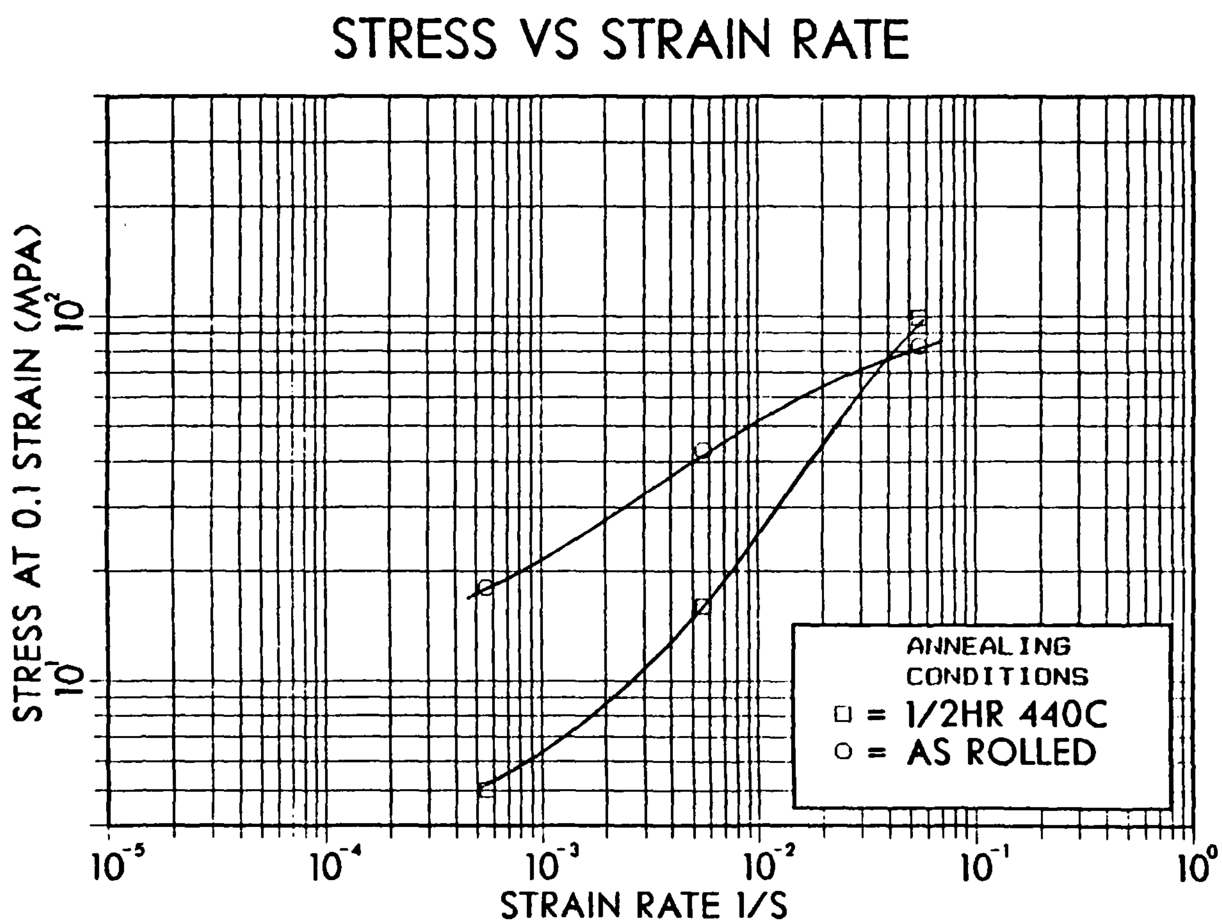


Figure A.21 True Stress at 0.1 Strain vs Strain Rate for Tensile Tests on Al-10.2%Mg-0.52%Mn Conducted at 350C. Data Indicates that Material Recrystallized by Annealing 1/2 Hour at 440C is Now Weaker than As-Rolled Material.

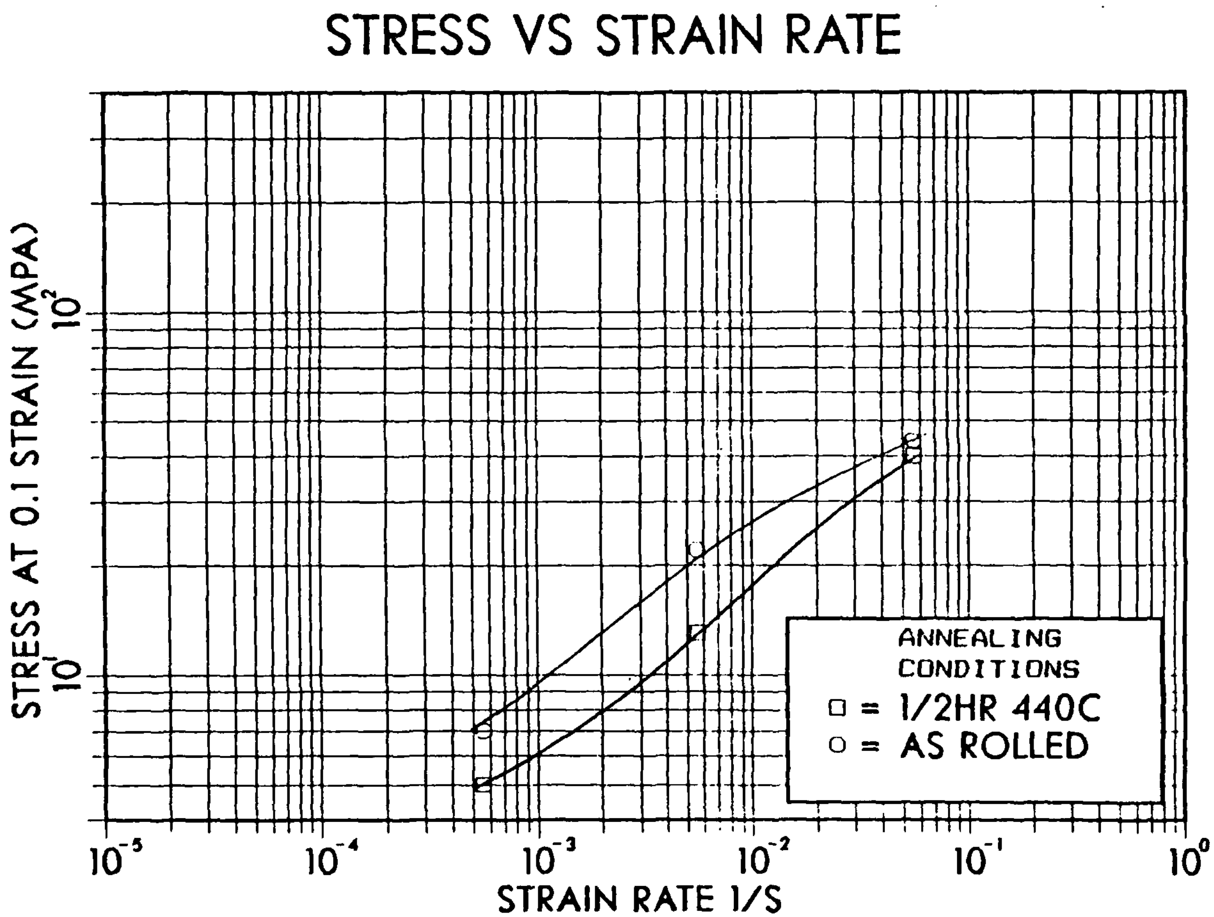


Figure A.22 True Stress at 0.1 Strain vs Strain Rate for Tensile Tests on Al-10.2%Mg-0.52%Mn Conducted at 425C. As Observed for Tensile Testing at 350C the Material Recrystallized by Annealing 1/2 Hour at 440C Continues to be Weaker then As-Rolled Material.

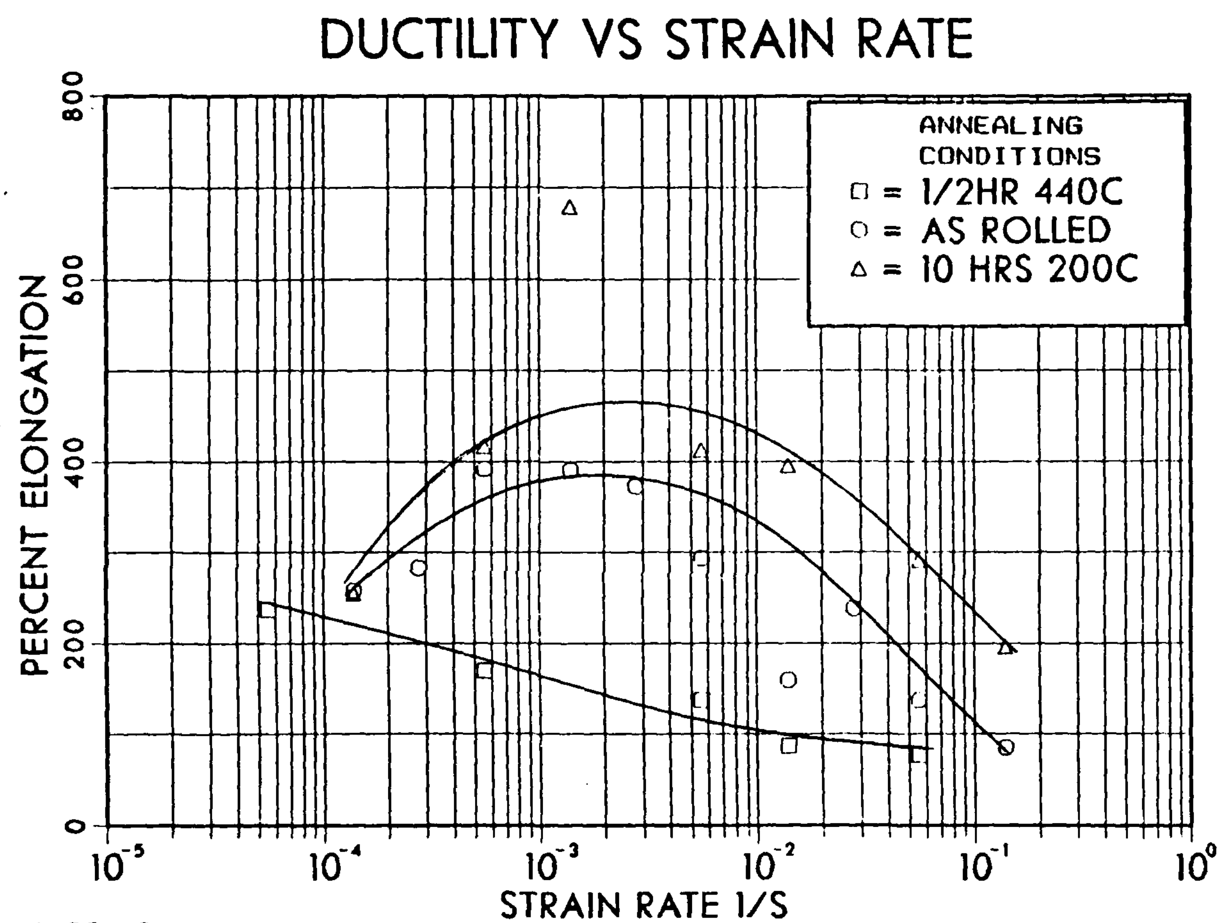


Figure A.23 Ductility vs Strain Rate for Tensile Tests on Al-10.2%Mg-0.52%Mn Conducted at 300C. Stress-Strain Rate Data for These Conditions is in Fig. A.17. Material Recrystallized by Annealing 1/2 Hour at 440C is Stronger and of Lower Ductility than As-Rolled Material at 300C. Annealing As-Rolled Material Below the Rolling Temperature Weakens the Material and Enhances the Superplastic Elongation.

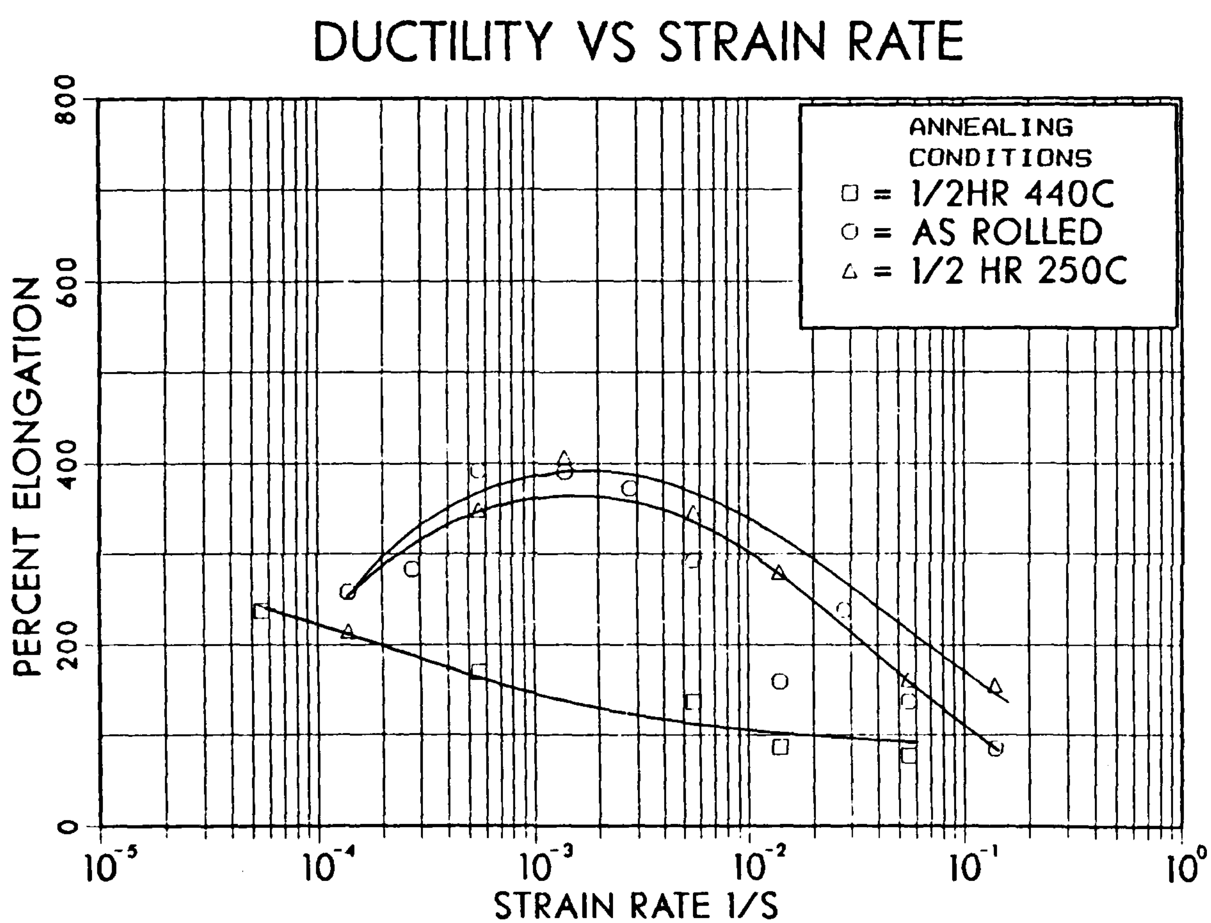


Figure A.24 Ductility vs Strain Rate for Tensile Tests on Al-10.2%Mg-0.52%Mn Conducted at 300C. Stress-Strain Rate Data for These Conditions is in Fig. A.18. Material Recrystallized by Annealing 1/2 Hour at 440C is Stronger and of Lower Ductility than As-Rolled Material at 300C. Annealing As-Rolled Material Below the Rolling Temperature Weakens the Material and Enhances the Superplastic Elongation.

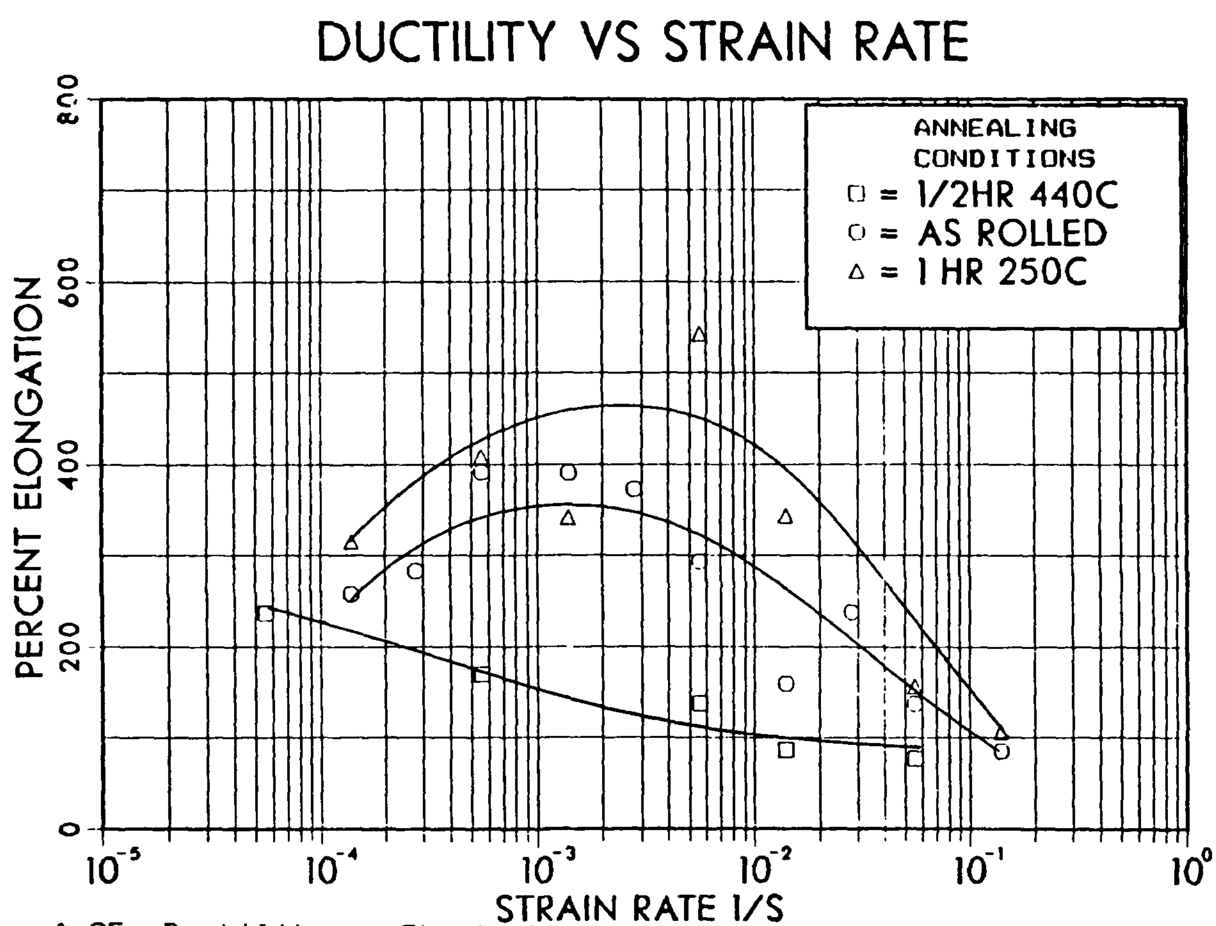


Figure A.25 Ductility vs Strain Rate for Tensile Tests on Al-10.2%Mg-0.52%Mn Conducted at 300C. Stress-Strain Rate Data for These Conditions is in Fig. A.19. Material Recrystallized by Annealing 1/2 Hour at 440C is Stronger and of Lower Ductility than As-Rolled Material at 300C. Annealing As-Rolled Material Below the Rolling Temperature Weakens the Material and Enhances the Superplastic Elongation.

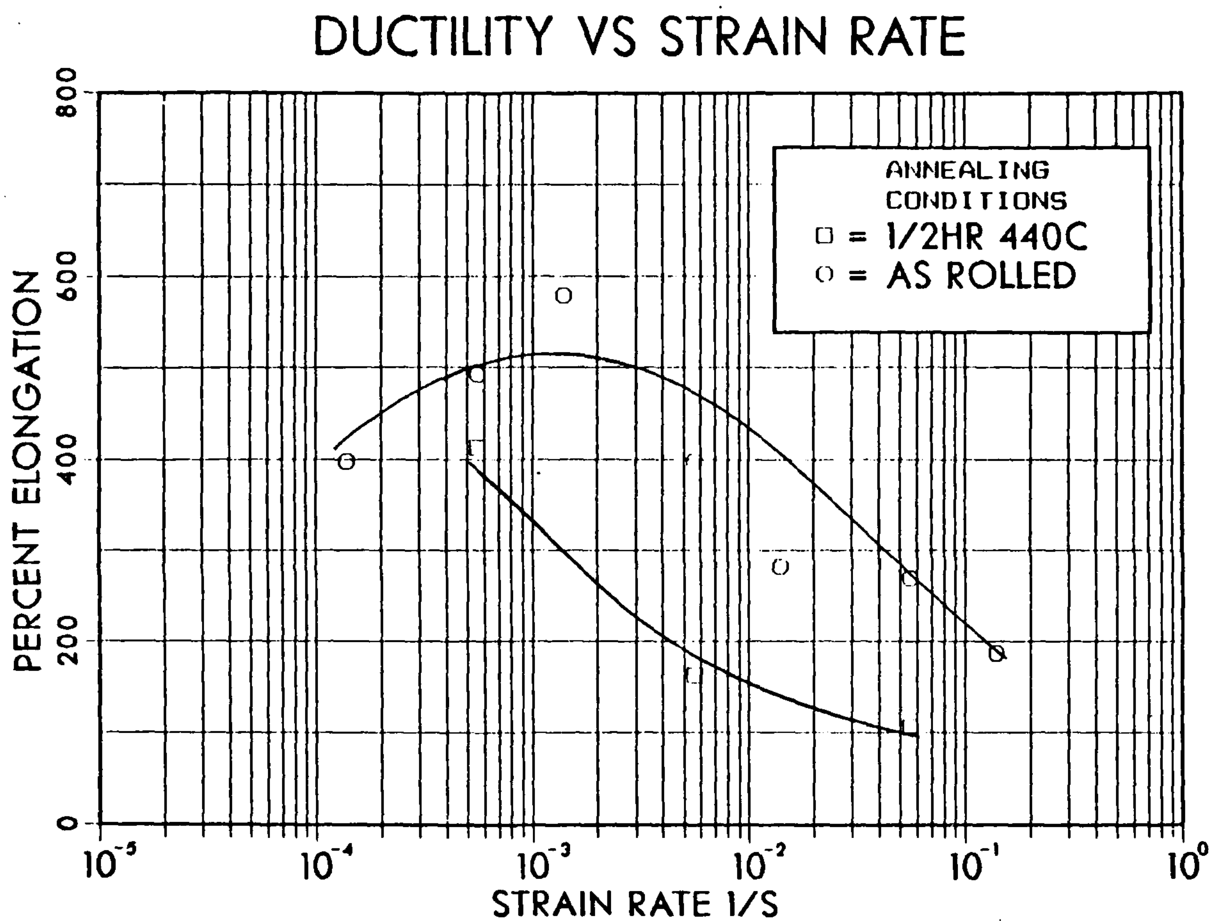


Figure A.26 Ductility vs Strain Rate for Tensile Tests on Al-10.2%Mg-0.52%Mn Conducted at 325C. Annealing Conditions: Fully Recrystallized after Warm Rolling at 300C, and As Rolled at 300C.

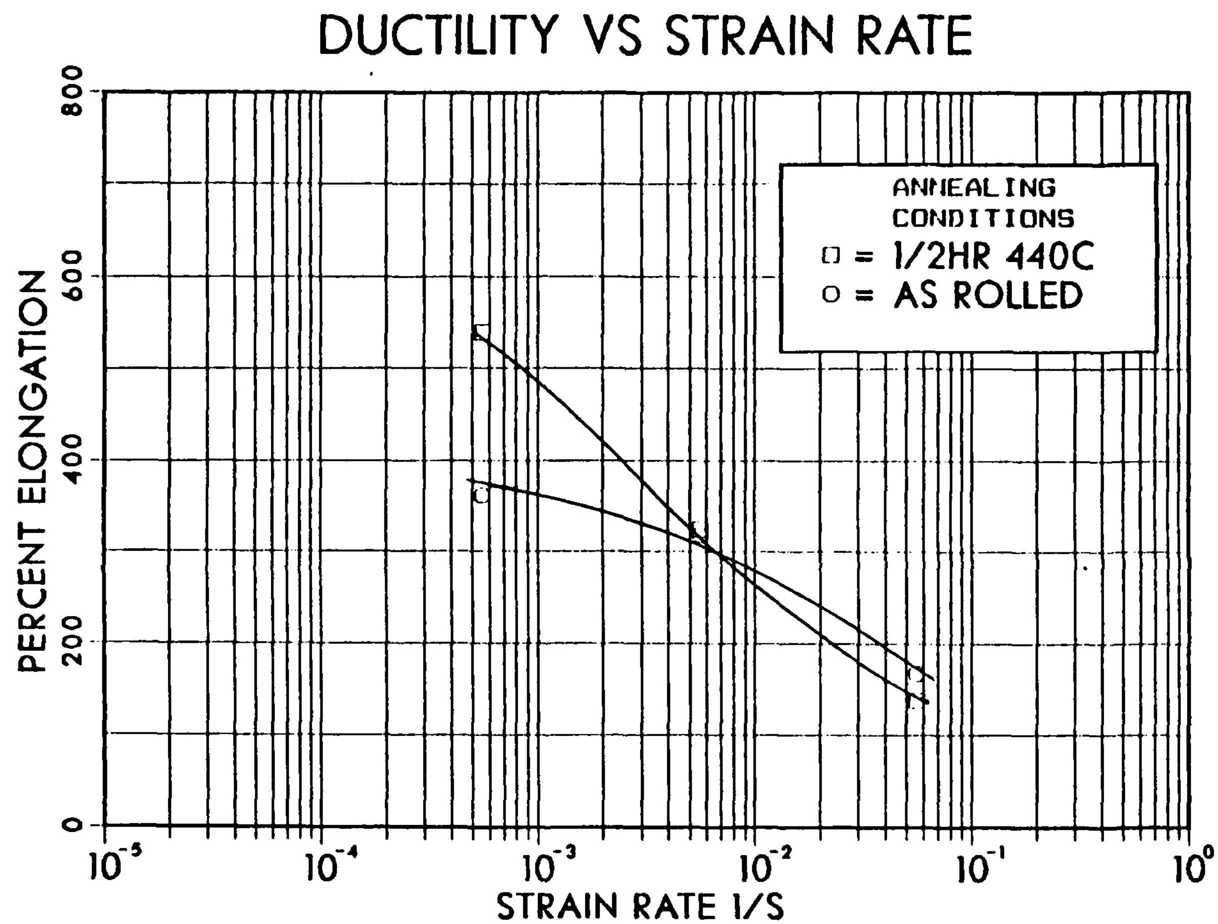


Figure A.27 Ductility vs Strain Rate for Tensile Tests on Al-10.2%Mg-0.52%Mn Conducted at 350C. Annealing Conditions: Fully Recrystallized after Warm Rolling at 300C, and As Rolled at 300C.

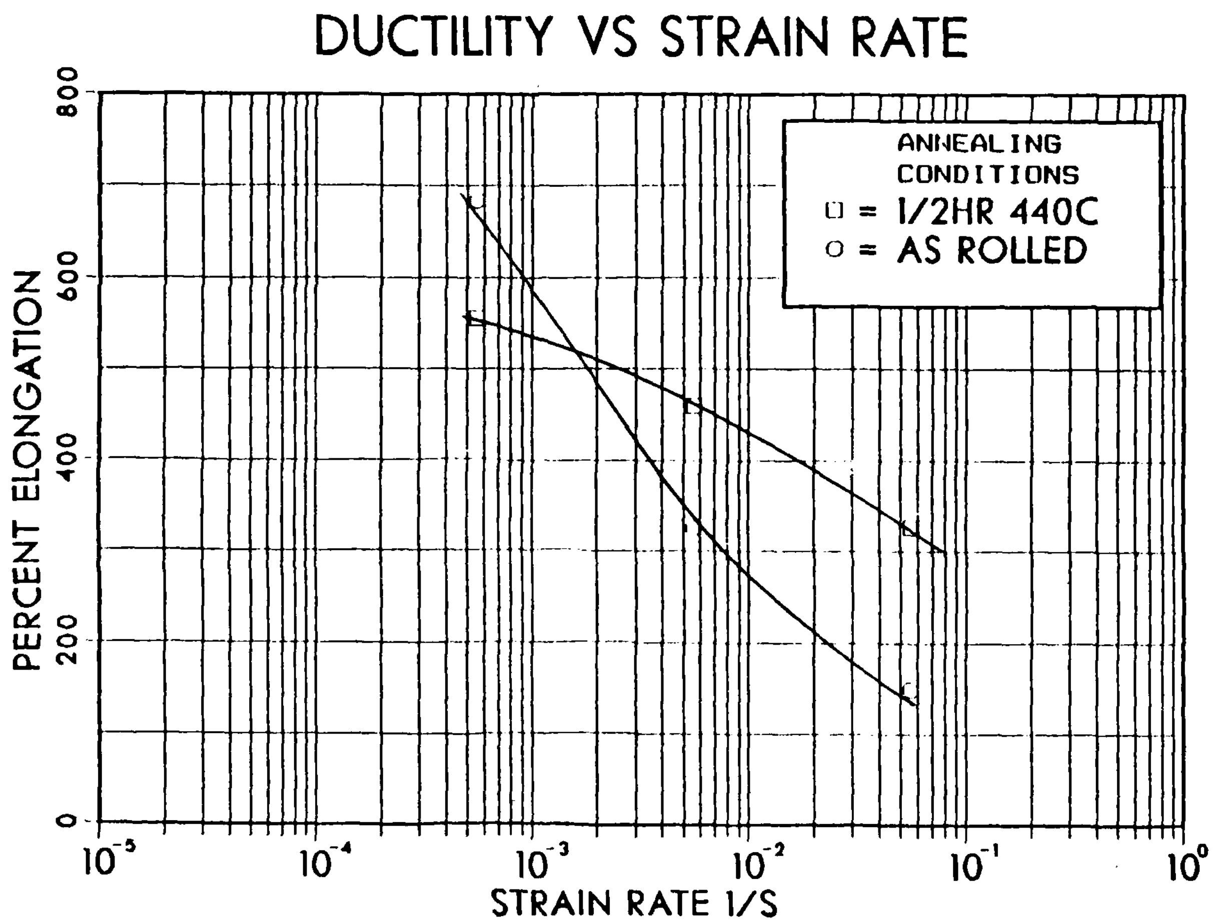


Figure A.28 Ductility vs Strain Rate for Tensile Tests on Al-10.2%Mg-0.52%Mn Conducted at 425C. Annealing Conditions: Fully Recrystallized after Warm Rolling at 300C, and As Rolled at 300C.

APPENDIX B

```
10 DIM X(15),Y(15),EENG(55,25),SEMPA(55,25),ETRUE(55,25),STMPA(55,25)
20 FOR J = 1 TO 41
30 PRINT"File number",J
40   READ TEMP$,PE$,S$,W,T,M,R,A$
50   RR=R/36!
60   REM PRINT HEADING
70   REM GOSUB 290
80   FOR I = 1 TO 11
90     READ X(I),Y(I)
100     A=W*T
110     SENG=(X(I)*500!)/(10!*A)
120     EENG(J,I)=Y(I)/(M*.6)
130     SEMPA(J,I)=SENG*.006895
140     STRUE=SENG*(1+EENG(J,I))
150     STMPA(J,I)=STRUE*.006895
160     ETRUE(J,I)=LOG(1+EENG(J,I))
170     REM PRINT DATA
180     REM GOSUB 410
190   NEXT I
200       BB$=STR$(J)
201 IF LEFT$(BB$,1)=" " AND J>9 THEN BB$=RIGHT$(BB$,2) ELSE BB$=RIGHT$(BB$,1)
210       AA$="c:file"+BB$+".dat"
220       OPEN AA$ FOR OUTPUT AS #1
230       PRINT AA$
240   FOR I = 1 TO 11
250       PRINT USING "  #.######^^^^";EENG(J,I),SEMPA(J,I),ETRUE(J,I),STMPA(J,I
)
260      PRINT #1,USING "  #.######^^^^";EENG(J,I),SEMPA(J,I),ETRUE(J,I),STMPA(J
,I)
270   NEXT I
280   CLOSE
290   NEXT J
300 END
```

```
310 REM SUBROUTINE TO PRINT HEADINGS
320   PRINT "***************************************************************************
***********"
330   PRINT "    W    ";"    T    ";"    M    ";"  E RATE";"    ANNEAL";"    SAM
PLE";"     TEMP   ";" % ELONG"
340   PRINT USING "###.#### ";W,T,
350   PRINT USING "  ###.   ";M,
360   PRINT USING " ##.##^^^^";RR,
370   PRINT USING "      !";A$,
380   PRINT USING "        \      \";S$,
390   PRINT USING "   \     \";TEMP$,PE$
400   PRINT "   Cp";"      Cx    ";"   SENG    ";"SENG    ";" EENG   ";" STRUE   ";"S
TRUE  ";" ETRUE"
410   PRINT "     ";"            ";"    PSI   ";" MPA    ";"         ";"  PSI    ";"
 MPA       "
420 RETURN
430 REM SUBROUTINE TO PRINT DATA
440     PRINT USING "###.##  ";X(I),Y(I),
450     PRINT USING "######.#";SENG,
460     PRINT USING "####.# ";SEMPA,
470     PRINT USING "##.#### ";EENG,
480     PRINT USING "######.#";STRUE,
490     PRINT USING " ###.# ";STMPA,
500     PRINT USING "##.#### ";ETRUE
510 RETURN
520 DATA 300C,107,4101-1,.1225,.0701,10,5.0,D,2.40,0.0,2.60,.04,2.90,.15,3.10,.2
9,3.14,.49,3.10,.70,2.70,1.58,1.80,3.55,.90,5.45,.50,6.11,0.0,6.80
```

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